PROCEEDINGS OF

THE INSTITUTE

Reports of Council · Meetings · Notices

Report of Council for 1956

THE COUNCIL submit this, their Annual Report and Statement of Accounts for the year 1956, to Members for their approval at the 88th Annual General Meeting of The Iron and Steel Institute.

The Annual General Meeting was held in May in London and was associated with the celebration of the Centenary of the Bessemer Process; the Autumn General Meeting was also held in London in November, and there was a Special Meeting in France in June. The Engineers Group met twice in London and once in South Wales. A Joint Meeting on Corrosion was held with B.I.S.R.A. in London in October. A number of Joint Meetings were organized with the Affiliated Local Societies.

ROLL OF THE INSTITUTE

The total membership at 31st December was 5084 compared with 5002 at 31st December, 1955.

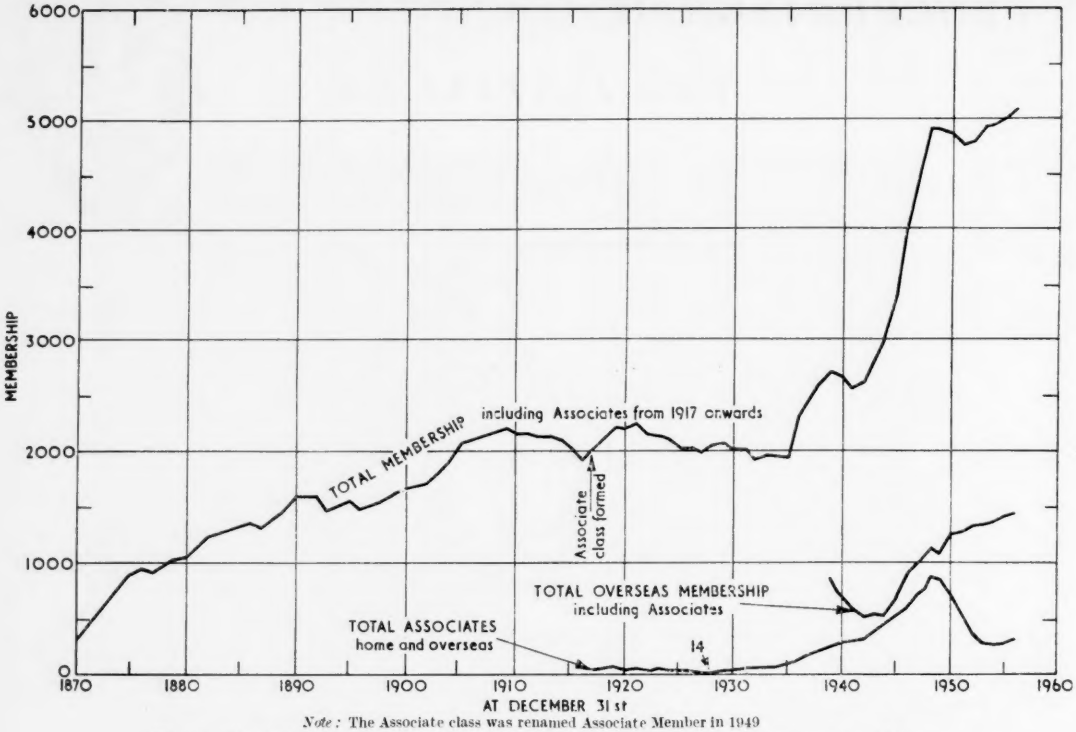
During the year, 241 Ordinary and 95 Associate Members were elected or reinstated. The deaths of 49 were reported: 152 resigned, 44 were struck off the Roll of Members, and 9 elections lapsed for non-payment of subscriptions.

An analysis of the membership at 31st December, 1956, is given in Table I, and the membership trend since the Institute was founded is shown in Fig. 1.

OBITUARY

The Council regret to announce the deaths of the following 49 Members which were reported during 1956:

ADAMSON, J. L. (Ewell)	7th February, 1956
ALLEN, Sir Richard (Bedford)	July, 1955
BACON, C. H. (Chester)	26th August, 1956
BINNIE, Dr. D. (Manchester)	3rd October, 1956
BOOT, H. (Consett)	8th February, 1956
BOWEN, W. T. (Worcester)	7th February, 1956
BROWN, M. (Henley-on-Thames)	14th November, 1956
BUSSEY, J. (Sheffield)	15th June, 1956
COCKIN, E. (Corby)	June, 1955
DALE, James (Larkhall)	31st January, 1956
DAVISON, V. W. (address unknown)	28th April, 1956
DUNN, Sir James (Montreal)	1st January, 1956
ELSDON, R. (Sutton)	21st February, 1956
GEIB, M. (Denain)	23rd February, 1956
GILFILLAN, G. (Sheffield)	17th January, 1956
GRAHAM, G. M. (Jorpeland, Norway)	21st June, 1956
GRAY, B. (Sheffield)	7th October, 1956
GREEN, G. W. (Leeds)	October, 1955
HILL, E. G. (Wheeling, West Virginia, U.S.A.)	July, 1956
HOLLINGS, J. S. (Cardiff)	21st June, 1956
HOUGHTON, E. (Chesterfield)	30th October, 1956
JONES, H. S.	21st January, 1956
KERL, E. (Pretoria, South Africa)	21st September, 1956
KERR, J. SINCLAIR (Manchester)	14th December, 1956
LEPP, H. (Paris, France)	25th May, 1956
LILLIEQVIST, G. A. (East Chicago, Ind., U.S.A.)	31st May, 1955
LLOYD, E. C. (Cwmfelinfach)	Not known
LONG, H. M. (Montreal, Canada)	20th February, 1956
LOWE, E. J. (Sheffield)	10th August, 1956
MCKEE, A. G. (Cleveland, Ohio, U.S.A.)	19th February, 1956
MCNEIL, J. (Fife)	21st February, 1956
MENZIES, K. (London)	6th June, 1956
MONTGOMERY, J. (Motherwell)	April, 1956
PATWARDHAN, M. K. (Bombay)	18th October, 1955
PAUL, R. P. J. (London)	12th January, 1956
PEARSON, S. W. (Rotherham)	21st January, 1956
PICKERING, W. B. (Sheffield)	24th April, 1955
RAMSEYER, C. F. (New York, N.Y., U.S.A.)	9th December, 1956
RAYNER, P. J. (Sheffield)	5th September, 1955
RIDSDALE, N. D. (Middlesbrough)	17th July, 1956
ROSEVERE, G. R. (Torquay)	7th June, 1956
SKERL, J. G. A. (St. Albans)	31st January, 1956
SHILLITOE, H. (London)	6th September, 1956
SMITH, W. C. (Pentyrch, Glam.)	3rd March, 1956
SUESS, Dr. T. E. (Kochel am See, Germany)	4th March, 1956
WESTBERG, S. (Rotherham)	Not known



Note: The Associate class was renamed Associate Member in 1949

Fig. 1—Variation in Membership since the foundation of the Institute in 1869

Table I

MEMBERSHIP OF THE IRON AND STEEL INSTITUTE AT 31st DECEMBER, 1956, AND COMPARISON WITH 1955

	Membership at 31st December, 1956													1955		
	Ordinary Membership		Joint Membership with								Total Membership		Grand Totals	Total Membership		Grand Totals
			Institute of Metals		Institution of Metallurgists		Both Institute and Institution		Totals							
	Home	Overseas	Home	Overseas	Home	Overseas	Home	Overseas	Home	Overseas	Home	Overseas		Home	Overseas	
Patron	1	...	...	...	...	...	...	...	...	...	1	...	1	1	...	1
Members:																
Hon. Vice-Presidents	6	9	1	1	...	...	...	...	1	1	7	10	17	8	8	16
Hon. Members	3	7	1	1	...	...	...	1	1	2	4	9	13	4	7	11
Non-paying Members	5	1	...	...	...	...	...	...	...	...	5	1	6	5	1	6
Life Members	59	35	3	6	8	...	6	...	17	6	76	41	117	79	...	117
Ordinary Members	1842	871	440	365	541	53	444	62	1425	480	3267	1351	4618	3251	1321	4572
Total (Members)	1916	923	445	373	549	53	450	63	1444	489	3360	1412	4772	3348	1375	4723
Associate Members	62	17	190	15	20	2	6	...	216	17	278	34	312	247	32	279
Total (both classes)	1978	940	635	388	569	55	456	63	1660	506	3638	1446	5084	3595	1407	5002

WHITFIELD, R. (Smethwick, Staffs.)	17th June, 1956
WINGATE, D. (Glasgow)	5th September, 1956
WYLIE, D. C. (Motherwell)	2nd January, 1956

Mr. J. S. Hollings, C.B.E., had been a Member since 1903 and was an Honorary Vice-President; he was awarded the Bessemer Gold Medal in 1946 and was elected an Honorary Member at the Annual General Meeting in 1956. Mr. J. Sinclair Kerr, a Member since 1911, was a Vice-President of the Institute, and Mr. Harold Boot was serving as a Member of Council at the time of his death. Another Honorary Member who died during the year was Mr. Richard Elsdon, Librarian of the Institute from 1908 until his retirement in 1951. Mr. Harold Shillitoe had been a Member since 1914.

HONORARY MEMBERSHIP OF THE INSTITUTE

The Council elected the following to be Honorary Members of the Institute during the year:

Sir John Craig, C.B.E. (*Past-President*)

Mr. J. S. Hollings, C.B.E. (*Honorary Vice-President*)

Monsieur Henri Malcor (*President of the Institut de Recherches de la Sidérurgie*)

Monsieur Jean Raty (*President of the Chambre Syndicale de la Sidérurgie Française*)

CHANGES ON THE COUNCIL

President-Elect

It was announced at the Autumn General Meeting, 1956, that the Council had unanimously decided to nominate Mr. A. H. Ingen-Housz, President of the Royal Netherlands Blast Furnaces and Steelworks Ltd., for election as President at the Annual General Meeting 1957, to hold office for one year. Mr. Ingen-Housz has been a Member since 1931 and was elected an Honorary Vice-President in 1953. This is the first time since 1918 that an Overseas Member of the Institute has been nominated for election as President.

Other Changes

During 1956 Dr. E. L. Dupuy and Baron Walckenaer were elected Honorary Vice-Presidents, and Sir Charles Goodeve, O.B.E., D.SC., F.R.S., was elected a Vice-President.

Major C. J. P. Ball, President of The Institute of Metals, agreed to serve as an Honorary Member of Council during his period of office, following Dr. Maurice Cook.

The following agreed to serve as Honorary Members of Council during their Presidencies of the Affiliated Local Societies named: Mr. H. Allsop (Sheffield Metallurgical Association, following Mr. T. H. Arnold); Mr. L. K. Everitt (Sheffield Society of Engineers and Metallurgists, following Mr. R. E. S. Fisher, M.C.); Mr. F. B. George (Cleveland Institution of Engineers, following Mr. H. H. Utley); Mr. W. H. R. Bird (Ebbw Vale Metallurgical Society, following Mr. J. F. Smith); Mr. A. Lenton (Leeds Metallurgical Society, following Mr. C. Breckon); Mr. C. T. Thomas (Lincolnshire Iron and Steel Institute, following Lt. Cdr. G. W. Wells, R.N. (RETD.); Mr. W. Rose (Liverpool Metallurgical Society, following Mr. G. T. Callis); Mr. G. A. Cottell (Manchester Metallurgical Society, following Mr. C. J. Bushrod); Mr. A. Biddulph (Staffordshire Iron and Steel Institute, following Mr. K. G. Lewis); Mr. R. Walker (Swansea and District Metallurgical Society, following Mr. R. W. Evans). Mr. W. C. F. Hessenberg agreed to serve as an Honorary Member of Council, representing the British Iron and Steel Research Association, in place of Sir Charles Goodeve, O.B.E., D.SC., F.R.S., who was elected a Vice-President.

In accordance with Bye-Law 10, the following names of Vice-Presidents and Members of Council due to retire at the Annual General Meeting, 1957, were announced at the Autumn General Meeting, 1956:

Vice-Presidents: Mr. C. R. Wheeler, C.B.E., Mr. D. F. Campbell, and Mr. G. H. Latham.

Members of Council: Mr. S. Thomson, Mr. E. T. Judge, Mr. E. J. Pode, Mr. B. Chetwynd Talbot, and Mr. W. L. James.

No other Members having been nominated up to one month before the Annual General Meeting, the retiring Vice-Presidents and other Members of Council named above are presented for re-election.

AWARDS AND PRIZES

The *Bessemer Gold Medal* for 1956 was awarded to Dr. C. Sykes, F.R.S., of Thos. Firth and John Brown Ltd., in recognition of his distinguished contribution to the study of alloy systems, particularly in the field of order–disorder changes and the effect of such changes on the physical properties of alloys.

The *Sir Robert Hadfield Medal* for 1956 was awarded to Mr. H. Morrogh of the British Cast Iron Research Association, in recognition of his distinguished contribution to the study of the metallography of cast iron and for researches which resulted in the discovery of nodular iron and the improvement in the quality of cast iron.

An additional *Williams Prize* for 1954 of £100 was awarded to be divided between Mr. F. A. Gray and Mr. S. H. Brooks (The United Steel Companies Ltd.) for the paper on "An Investigation of Reheating Furnace Design and Performance" (*Journal*, 1954, vol. 178, November), of which they were joint authors.

Two *Williams Prizes* for 1955 (value £50 each) were awarded to Dr. A. J. Kesterton (Steel Company of Wales Ltd.) for his paper on "Some Starting and Operating Experiences at Abbey Melting Shop" (*Journal*, 1955, vol. 179, January), which was presented before the Cleveland Institution of Engineers, and Mr. R. Wilcock (Samuel Fox and Co., Ltd.) for his paper on "The Operation of a 60-ton Arc Furnace at Samuel Fox and Co., Ltd." (*Journal*, 1955, vol. 180, June).

Three *Ablett Prizes* for 1955 (value £100 each) were awarded to Mr. W. Ernest (The United Steel Companies Ltd.) for his paper on "Waste Heat Recovery related to the Generation of Electric Power" (*Journal*, 1955, vol. 179, March), to Mr. S. R. Phelps (Guest, Keen and Nettlefolds (South Wales) Ltd.) for his paper on "Flying Shears for Bars and Billets: Electrical Requirements," (*Journal*, 1955, vol. 181, November), and Mr. R. Stewartson (Guest, Keen and Nettlefolds (South Wales) Ltd.) for his paper on "Flying Shears for Bars and Billets: Mechanical Design Features" (*Journal*, 1955, vol. 181, November).

ANDREW CARNEGIE RESEARCH GRANTS

During 1956 the following grants were made from the Andrew Carnegie Research Fund:

J. W. BARTON (Liverpool University): £400 (second grant) for a research on the use of paraffin wax to simulate strain patterns during the plastic deformation of metal.

M. PRASAD (Liverpool University): £250 for a research on the peeling of malleable cast iron.

K. A. SALMON (Sheffield University): £400 for a research on the kinetics of the transfer of sulphur between iron and slag.

P. VASUDEVIN (Manchester University): £400 for a study of the progress of isothermal transformation by the Eggertz colour method.

MOND NICKEL FELLOWSHIPS

Mr. F. H. Saniter continued to represent the Institute on the Mond Nickel Fellowships Committee.

The Committee made the following awards for 1956:

R. BANDY (The English Steel Tool Corporation Ltd.) to study in the United Kingdom, Europe, and North America metallurgical research, production methods, and quality control with respect to tool steels.

Dr. J. HARGREAVES (The United Steel Companies Ltd.) to study the changes in design, construction, and the methods of operation of steelmaking and processing furnaces which are being introduced in the United Kingdom, Europe, and the United States to increase the rate and efficiency of production.

G. F. C. DU PLESSIS (South African Iron and Steel Industrial Corporation Ltd.) to study the preparation of ores by sintering, pelletizing, etc., and the effect on blast-furnace practice of the use of prepared burdens.

ANNUAL GENERAL MEETING, 1956

The Eighty-seventh Annual General Meeting was held in London on Wednesday and Thursday, 16th and 17th May. The morning session for Wednesday, 16th May, was held at the Royal Institution, Albemarle Street, London, W.1; the remaining sessions were held at the Offices of the Institute. After the Report of Council and Accounts for 1955 had been approved and other formal business transacted, Sir Charles Bruce-Gardner, BT., inducted Dr. H. H. Burton, C.B.E., into the Chair. The new President delivered his Presidential Address on "Problems and Progress in Manufacture and Metallurgy" (*Journal*, 1956, vol. 183, June, pp. 113–123). The Ninth Hatfield Memorial Lecture was then given by Professor N. F. Mott, F.R.S., on "Fracture in Metals" (*Journal*, 1956, vol. 183, July, pp. 233–243). The proceedings of the Meeting are recorded on pp. 345–348 of the *Journal* for August, 1956, and the report of the discussions on pp. 438–461 of the December, 1956, *Journal*.

A Dinner for Members was held at Grosvenor House, Park Lane, London, W.1, on Wednesday, 16th May, at which Lieut.-General Sir Ronald Weeks, K.C.B., C.B.E., D.S.O., M.C., T.D., Chairman of Vickers Ltd., was the principal guest.

BESSEMER CENTENARY CELEBRATIONS

Celebration of the 100th anniversary of the invention of the Bessemer process was associated with the Annual General Meeting. A lecture on "Sir Henry Bessemer 1813–1898" was given by Mr. James Mitchell, C.B.E., Past-President, at the Royal Institution, Albemarle Street, London, W.1, on the evening of Tuesday, 15th May. Afterwards a Reception was held at 4 Grosvenor Gardens for Members of Council and their wives and for overseas and other visitors. On the afternoon of the Annual General Meeting on Wednesday, 16th May, two papers on the Bessemer process were discussed and a film was shown of the development of the process at the works of the Workington Iron and Steel Co., Ltd. The text of the Bessemer lecture and of these papers and other papers specially provided for the Centenary were published in the June, 1956, *Journal*. An exhibition of portraits, photographs, medals, and other items associated with Sir Henry Bessemer, most of which had been kindly lent by the Bessemer family, was held at the offices of the Institute from 15th to 18th May. A list of the exhibits and details of other functions held in connection with the Bessemer Centenary are published in the October, 1956, *Journal*.

SPECIAL MEETING IN FRANCE

A Special Meeting in France was held from 5th to 14th June, by invitation of the Chambre Syndicale de la Sidérurgie Française and the Société Française de Métallurgie. The programme consisted of six days in Paris, followed by two alternative tours to Central France and to Lorraine. About three hundred and fifty Members and ladies took part. A short account of the meeting was published in the *Journal* in January, 1957. An Opening Session was held in the Grand Amphithéatre of the Sorbonne in Paris, which concluded with a lecture on "Sir Henry Bessemer" by Monsieur G. Delbart, Director of the Institut de Recherches de la Sidérurgie. The text is given on pp. 190–195 of the June, 1956, *Journal*. Two Technical Sessions were held in the mornings of Wednesday and Friday, 6th and 8th June. Papers presented at these Sessions are published in the May and June, 1956, *Journals* and a report of the discussions is published in the April, 1957, *Journal*. The Council wish to express their thanks to the President, Monsieur Jean Raty, and Council of the Chambre Syndicale de la Sidérurgie Française; to the President, Baron Walckenaer, and Council of the Société Française de Métallurgie; to the Chairmen, Vice-Chairmen, and Members of the Reception and Ladies Committees; to the President of the Chamber of Commerce of Paris; to the Chambre Syndicale de la Sidérurgie de la Moselle and the Union des Mines et de la Métallurgie de Longwy; to Dr. E. Dupuy, Past-President and Honorary General Secretary of the Société Française de Métallurgie, and to Monsieur P. Gouge, Honorary General Secretary of the Société Métallurgique de Normandie; to General A. Gruenther and Staff of Supreme Headquarters Allied Powers in Europe; to the managements of the works and other establishments in France who kindly invited Members and Ladies to visit them; to Madame Raty, Baronne Walckenaer, and the Ladies who assisted in entertaining Members and Ladies; and to the authors of papers and all others who collaborated in organizing the Meeting.

MEETING ON CORROSION

A one-day Meeting to discuss a group of papers on "The Corrosion of Iron and Steel" was held jointly with the British Iron and Steel Research Association at the Offices of the Institute on Friday, 12th October. Dr. H. H. Burton, C.B.E., President, was in the Chair and was supported by Dr. J. Pearson. A report of the discussion at the Meeting is published in the May, 1957, *Journal*.

AUTUMN GENERAL MEETING

The Autumn General Meeting was held at the Offices of the Institute on Wednesday and Thursday, 21st and 22nd November, with the President in the Chair. The

proceedings of the opening session are published in the *Journal* for February, 1957, and a report of the discussion is to be published in the *Journal* later in the year.

IRON AND STEEL ENGINEERS GROUP

Mr. W. M. Larke continued to serve as Chairman of the Iron and Steel Engineers Group. The Engineering Committee, which organizes the activities of the Iron and Steel Engineers Group, was composed of the following Members (from June, 1956):

Chairman
Mr. W. M. Larke

Members

Mr. W. H. Alvey	Mr. R. A. Lake
Mr. D. R. Brown	Mr. H. M. Morgan
Mr. R. Cameron	Mr. J. Morgan
Sir Henry Clay, bt.	Mr. C. E. H. Morris
Mr. W. A. J. Dinwoodie	Mr. D. E. Moxey
Mr. J. H. Duffey	Mr. G. A. Phipps
Mr. G. Foster	Mr. F. E. Probyn

Mr. C. M. Slocombe

Ex-Officio
Dr. H. H. Burton, c.b.e., Mr. W. Barr, o.b.e., Dr. H. R. Mills

Two full-day meetings, with buffet luncheons, were held in London during the year. The Group also held a two-day meeting in South Wales, by invitation of the Steel Company of Wales Ltd., when two papers were presented, namely, "The Engineering Aspects of a Modern Basic Bessemer Plant" and "Planning the Conversion of a High-Lift Slabbing to a Universal Mill." Visits were made to the Margam and Abbey Works of the Steel Company of Wales Ltd.; approximately 250 members attended.

PUBLICATIONS

The *Journal* was again published in twelve monthly parts, comprising Volumes 182, 183, and 184. Indexes and binding cases are being prepared.

Special Reports—Five Special Reports were published during the year: "Physical Aspects of Absorptiometric Analysis" (No. 55), "Atlas of Isothermal Transformation Diagrams" (No. 56), "The Russian Iron and Steel Industry" (No. 57), "Symposium on Powder Metallurgy, 1954" (No. 58), and "Combustion and Heat Transfer in an Open-Hearth Furnace" (No. 59).

Bibliographies—No Bibliographies were added to the series during the year.

Translation Service—Translations of eight foreign technical papers (Nos. 507–514) were issued during 1956. The charge was £1 each (10*s*. for each additional copy of the same translation). Nineteen translations were made for individual members which were not incorporated into the service.

Handbook and List of Members, 1956—A Handbook giving information on the constitution, history, and activities of the Institute and including a List of Members arranged alphabetically and topographically was again issued to all members during the year.

The British Welding Journal and the *Transactions of the Society of Instrument Technology*—The Institute continued the production of the *British Welding Journal* and the *Transactions of the Society of Instrument Technology*. The Institute of Welding, the British Welding Research Association, and the Society of Instrument Technology again expressed their satisfaction at the results achieved.

JOINT LIBRARY AND INFORMATION DEPARTMENT

During 1956, 12,690 publications were borrowed from the Joint Library, compared with 14,416 in 1955: the decrease is largely accounted for by the omission of publications taken out during 1956 for use by members of staff of The Iron and Steel Institute and The Institute of Metals. During the year 401 new text books were acquired (232 in 1955); the Council offer their thanks to all those who kindly presented copies of their books and papers to the Library.

Photographic copies of documents can, under certain conditions, be supplied to members, and during the year 211 photocopies and 45 microfilms were prepared (172 and 66 in 1955); of these about half of the photocopies and all the microfilms were sent overseas.

The services of the Information Department, like those of the Joint Library, were much in demand by members, Government Departments, Universities, and Research Departments; the Department seeks to answer enquiries by reference to the scientific and technical literature, and is careful to avoid encroaching on the field of the metallurgical consultant.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The close co-operation which has existed between the Institute and the British Iron and Steel Research Association ever since the formation of the latter was continued during the year. A detailed review of the services provided by the Institute to the Research Association was made during the year, with a view to introducing some extensions and improvements in the near future.

RELATIONS WITH OTHER SOCIETIES AND ORGANIZATIONS IN GREAT BRITAIN AND OVERSEAS

During the year the long-standing friendly relations between the Institute, The Institute of Metals, and The Institution of Metallurgists were maintained and strengthened. The friendliest co-operation also continued with the Iron and Steel Board, with the British Iron and Steel Federation, and also with the West of Scotland Iron and Steel Institute. Friendly relationships were also maintained with many scientific and technical societies at home and overseas.

Affiliated Local Societies in Great Britain

There are now twelve Affiliated Local Societies in Great Britain, as follows:

Cleveland Institution of Engineers
Ebbw Vale Metallurgical Society
Leeds Metallurgical Society
The Lincolnshire Iron and Steel Institute
Liverpool Metallurgical Society
The Manchester Metallurgical Society
The Newport and District Metallurgical Society
North Wales Metallurgical Society
Sheffield Metallurgical Association
Sheffield Society of Engineers and Metallurgists
Staffordshire Iron and Steel Institute
Swansea and District Metallurgical Society.

The arrangement for affiliation remained unchanged. The opinion of the Affiliated Societies continued to be to the effect that these arrangements were satisfactory and helpful to them. Their value to the Institute remained as high as in previous years.

Joint Meetings

During the year joint meetings were held between the Institute (including the Engineers Group) and the following Affiliated Local Societies: Ebbw Vale Metallurgical Society, Manchester Metallurgical Society, Newport and District Metallurgical Society, Sheffield Metallurgical Association, and the Sheffield Society of Engineers and Metallurgists. A joint meeting of the Engineers Group was also held with the West of Scotland Iron and Steel Institute.

Metallurgical Societies Overseas

The number of 'kindred' societies overseas remains at eighteen:

American Institute of Mining, Metallurgical, and Petroleum Engineers
American Iron and Steel Institute
American Society for Metals
Association des Ingénieurs sortis de l'Ecole de Liège (Belgium)
Associazione Italiana di Metallurgia (Italy)
Association Luxembourgeoise des Ingénieurs et Industriels (Luxembourg)
Australian Institute of Metals
Canadian Institute of Mining and Metallurgy
Indian Institute of Metals
Instituto del Hierro y del Acero (Spain)
Jernkontoret (Sweden)
Koninklijk Institut van Ingenieurs (Holland)
Norsk Metallurgisk Selskap (Norway)
Société Française de Métallurgie (France)
Société Royale Belge des Ingénieurs et des Industriels (Belgium)
Verein deutscher Eisenhüttenleute (Germany)
Verein Eisenhütte Oesterreich (Austria)
Verein Schweizerischer Maschinen-Industrieller (Switzerland).

Members of the Institute who are Members of, or are on the staffs of member firms of any of the above Societies, if resident in the countries concerned, pay reduced membership subscriptions to the Institute.

Representation at Meetings

Mr. W. F. Cartwright, Vice-President, represented the Institute at the Latin American Meeting of Experts on Steel Making and Transforming Industries at Sao Paulo, Brazil, from 15th to 28th October, 1956, which was organized by the United Nations Economic Commission for Latin America. Mr. K. Headlam-Morley, the Secretary, represented the Institute at the Meetings of the Société Française de Métallurgie in Paris in October and the Verein deutscher Eisenhüttenleute, Düsseldorf, in November, 1956. In Düsseldorf he presented to the Verein on behalf of the Council and Members of the Institute, in recognition of the 75th Anniversary of the publication of "*Stahl und Eisen*," the Bronze Medal and Certificate of Honorary Membership of the Verein zur Beförderung des Gewerbfleisses in Preussen contained in a wooden case with finely ornamented bronze cover which was presented to Sir Henry Bessemer, F.R.S., in 1877. Mr. A. E. Chattin, Assistant Secretary and Librarian, represented the Institute at the Annual Meeting of the Verein Eisenhütte Oesterreich in Leoben, Styria, Austria, on 2nd June, when he delivered a paper on "Sir Henry Bessemer, F.R.S."

PRESENTATIONS TO THE INSTITUTE

Mr. H. Douglas Bessemer and Miss R. E. M. Bessemer kindly presented a number of items that were the property of their father, the late Mr. Henry W. Bessemer, who was a grandson of Sir Henry Bessemer, F.R.S. Details of these items and of a number of other items loaned to the Institute by Mr. H. Douglas Bessemer and Miss R. E. M. Bessemer are given on pp. 118–120 of the October, 1956, *Journal*.

Imperial Chemical Industries Ltd., through The Institute of Metals, made a third gift of £21, and Gillette Industries Ltd. donated £105; both sums were given in appreciation of services rendered to their Companies by the Joint Library.

The Joint Library also benefited by gifts from Dr. E. Houdremont, Professor Cyril Smith, and Mr. G. V. Uzhik of books of which they were the authors, and from Dr. C. H. Desch, F.R.S., Mr. R. Mather, the American Embassy Library, the Chemical Society, the Institution of Mining and Metallurgy, and Welding Supplies Ltd. of a number of books on a variety of aspects of interest to Members of the Institute.

The Council offer their thanks to these donors for their generous gifts.

EDUCATION

Joint Committee on National Certificates in Metallurgy

The Institute was represented on the Joint Committee for National Certificates in Metallurgy by Mr. K. Headlam-Morley.

Sheffield University—Institute Prize, 1956

The Iron and Steel Institute Prize for 1956 at Sheffield University was awarded to Mr. G. A. Chadwick.

OFFICES OF THE INSTITUTE

The Institute of Metals moved in June, 1956, to new premises which they had acquired at 17 Belgrave Square, London, S.W.1. The Council are pleased to record that the close co-operation which has existed for so many years with The Institute of Metals will in no way be affected by their move to other premises. The arrangements for the Joint Library with The Institute of Metals remain unchanged. The offices on the fourth floor at 4 Grosvenor Gardens, previously occupied by The Institute of Metals, have been taken over on a seven-year lease by The Institution of Metallurgists. The Council wish to express their pleasure that The Institution of Metallurgists should again be accommodated in the building.

Three rooms at the back of the third floor have been let on short lease.

STAFF

The Council again wish to record their appreciation of the services rendered by the staff during the year.

HONOURS CONFERRED UPON MEMBERS

The Council offer their congratulations to all members who have received Honours or appointments during the year. The names have been recorded in the News section of the *Journal*.

FINANCE

The Accounts for 1956 are appended.

(Figures for the previous year are printed in brackets for comparison.)

GENERAL

Nearly 70% of the Institute's income during 1956 was obtained as the result of publishing activities and less than 20% from subscriptions paid by Members. This is

not necessarily a disadvantage. In fact, it is owing to the income obtained from other sources that it has been possible for Members' subscriptions to be kept at the present relatively low level. It is clear, however, that any change in the income from publishing—even a slight falling off in the tendency to increase that has been a marked feature of recent years—will have a correspondingly greater effect on the financial position as a whole. The position is being carefully watched.

The standard rate of subscription payable by an Ordinary Member was last increased in 1949 from £3 3*s*. 0*d*. to £4 4*s*. 0*d*. per annum and the average rate paid by Ordinary and Associate Members during 1956—Life, Honorary and Non-Paying Members excluded—was £3 12*s*. 2*d*. The Council's policy is to keep subscription rates down and no increase is needed at present. The position will, however, be reviewed from time to time and if necessary a recommendation will be made to Members. In view of constantly increasing prices, this possibility must be seriously considered.

Income during 1956 again exceeded expenditure by a small amount but a larger margin is required if transfers to Reserves for special foreseeable expenditure are to be made without drawing on the Accumulated Fund. The House Fund, founded in 1953, by subscriptions from companies in the industry, has been further increased.

GENERAL FUND

Balance Sheet and Income and Expenditure Account

The Balance Sheet at 31st December, 1956, and the Income and Expenditure Account for the year ending on that date are submitted in the usual form.

Ordinary income, at £87,646 (£79,982) exceeded ordinary expenditure, £86,040 (£77,244) by £1,606 (£2,738). After charging certain special items, the sum of £407, excess of income, was transferred to the Accumulated Fund. (In the previous year there was a deficit of £4,904, as a result of the expenditure on the Institute's building and other special items). Transfers of £1,750 were made from the Accumulated Fund to Reserves and the balance, £28,939 (£30,115) carried to the Balance Sheet.

A further £1,616 has been written off the value of the lease of No. 4 Grosvenor Gardens, with the result that the lease, which cost, including legal expenses, £43,518 in 1953, has been written down by £5,665 and now stands in the Balance Sheet at £37,853. The total value of the House Fund is £55,950 (£55,200).

TRUST FUNDS

The accounts of the Andrew Carnegie Research, Williams Prize, Bessemer Medal and Sir Robert Hadfield Medal funds are presented as usual. Full use has again been made of the available income. The following is a summary: Income £1,675, Expenditure £1,846, Accumulated Funds £7,874 (£8,045).

INVESTMENTS

The market value of investments at the end of the year reflected the general fall in the value of gilt-edged securities. Most of the stocks in which the Institute's Funds are invested have fixed redemption values and dates. Some changes have been made in the investments of the General and Trust Funds.

The investments are valued in the Balance Sheets at what they cost when purchased. In the case of the General, Life Composition and House Funds, the cost was £48,814 (£51,303) and their market value at 31st December, 1956 was £40,055 (£47,143). In the case of the four Trust Funds the cost was £38,683* (£40,678) and their market value at 31st December, 1956, was £34,666 (£39,303).

* This figure includes £474 15*s*. 0*d*. British Transport 3% Guaranteed Stock 1978-88 at the figure of £400, at which it is included in the Balance Sheet of the Bessemer Medal Fund.

THE IRON AND STEEL INSTITUTE

BALANCE SHEET, 31st DECEMBER, 1956

General Account

1955	LIABILITIES	£	£
	Accumulated Fund:		
30,115	Balance per Account annexed		28,939
	Current Liabilities and Provisions:		
12,541	Sundry Creditors (part estimated)	14,520	
1,056	Due to Life Composition Fund	2,619	
2,030	Due to House Fund	1,616	
35	Due to Sir Robert Hadfield Medal Fund	39	
26	Due to Bessemer Medal Fund	34	
1	Due to Williams Prize Fund	—	
—	Due to Andrew Carnegie Research Fund	14	
242	Ablett Prize Fund Account	43	
476	Subscriptions received in advance	316	
4,250	Payments received in advance for Publications	3,596	
	Sundry Provisions as at 31st December, 1956:		
	Reserve for Comprehensive Index	1,850	
	Transfer from Accumulated Fund	250	
		2,100	
	Less Expenditure during year	250	
1,850			1,850
	Reserve for Supplementing Staff Pensions	2,000	
	Transfer from Accumulated Fund	500	
2,000			2,500
			4,350
2,836	Bank Overdraft		—
27,343			27,147
£57,458			£56,086

1955	ASSETS	£	£
	Furniture and Fittings		Not valued
	Investments:		
	Balance at 31st December, 1955 at cost	32,983	
	Less Transfer to House Fund	2,030	
	,, Realizations during year	1,000	
		3,030	
32,982			29,953*
	(*Market value at 31st December, 1956—£24,884)		
	Current Assets:		
2,801	Stock of Paper as valued by the Managing Editor		2,050
8,251	Sundry Debtors		13,489
—	British Iron and Steel Research Association		326
5,160	British Welding Journal		4,353
1,091	Transactions of the Society of Instrument Technology		902
133	Current Accounts		—
804	Deposits and Payments in Advance		1,159
179	Balance at Banks and Cash in Hand		655
18,419			22,934
	Publishing Account:		
5,466	Expenditure incurred in advance on Publications		2,405
591	*Special Publications Suspense Account*		794
£57,458			£56,086

Special Funds

1955		£	£
	Life Composition Funds:		
	Balance at 31st December, 1955	5,995	
	Add Compounded during 1956	63	
	Premium on Conversion of Defence Bonds	10	
5,995			6,068
	House Fund:		
	Amounts received from Companies	55,200	
	Amount receivable in 1956	750	
55,200			55,950
£118,653			£118,104

1955		£	£
	Life Composition Funds:		
	Investments:		
4,939	Balance at 31st December, 1955 at cost	4,939*	
	Less Realizations during year	1,490	
			3,449*
1,056	Due from General Account		2,619
5,995	(* Market value at 31st December, 1956—£3,042)		6,068
	House Fund:		
39,469	Lease at Cost less amount written off, per Account annexed		37,853
319	Sundry Debtors		1,069
13,382	Investment at Cost		
	Balance at 31st December, 1955	13,382*	
	Add Transfer from General Fund	2,030	
	(* Market value at 31st December, 1956—£12,129)		15,412
2,030	Due from General Fund		1,616
55,200			55,950
£118,653			£118,104

We have examined the above Balance Sheet of the Institute and found it to be correct.

FINSBURY CIRCUS HOUSE,
BLOMFIELD STREET,
LONDON, E.C.2
29th March, 1957

(*Signed*) W. B. KEEN & Co.,
Chartered Accountants

INCOME AND EXPENDITURE ACCOUNT FOR THE YEAR ENDED 31ST DECEMBER, 1956

EXPENDITURE

1955 £	£		£	£
		Staff Expenses:		
27,775		Salaries	30,190	
521		National Insurance	590	
2,759		Staff Superannuation Fund and Pension	2,844	
932		Cost of Staff Luncheon Vouchers	985	
31,987				34,609
		Other Administrative Expenses:		
1,957		Printing and Stationery	2,328	
1,030		Postage and Receipt Stamps	1,150	
1,833		Travelling and Entertainment	2,300	
672		Meetings	726	
734		Telephone and Telegrams	770	
133		Insurances	224	
188		Audit Fee—1955	189	
603		Sundry Payments	589	
7,150				8,276
		Establishment Expenses:		
1,000		Rent	400	
742		Rates and Water	1,438	
1,906		Cleaning, Heating and Lighting	2,148	
437		Repairs and Decorations	847	
442		Office Furniture and Fittings	466	
4,527				5,299
		Publishing:		
26,659		1956 *Journal*—Printing, Paper, etc.	31,194	
3,012		Other Publishing Activities—Expenses	2,885	
78		Adjustments—amounts under-provided in previous year	—	
29,749				34,079
—		*Translations Service*—Preparation		219
		Library and Information Department:		
600		Library Books and Binding	781	
47		Reviewers' Fees	53	
935		Abstractors' Fees	990	
1,582				1,824
		Miscellaneous:		
1,318		Historical Investigator—Honorarium and Expenses	1,203	
256		Contributions to Affiliated Local Societies	211	
150		Joint Committee for National Certificates in Metallurgy	150	
10		Prizes	10	
294		Grants	160	
2,028				1,734
77,244				86,040
2,738		Balance carried down		1,606
£79,982				£87,646
—		Deficit on Meeting in France		269
—		Special Expenditure on Office Equipment		459
—		Cost of Bessemer Centenary Celebrations		829
2,099		Cost of Special Entertainment of Overseas Visitors		—
5,359		Expenditure on Building Improvements		—
566		Renovation of Council Room, Furniture and Pictures		—
—		Excess of Income over Expenditure for the year		407
£8,024				£1,964

INCOME

1955 £	£		£	£
		Subscriptions:		
17,072		Current	17,355	
527		Arrears	514	
17,599				17,869
		Payments for Services:		
5,000		British Iron and Steel Research Association	5,000	
150		Andrew Carnegie Research Fund	150	
250		Mond Nickel Fellowships Committee	250	
100		Institute of Metals	50	
5,500				5,450
1,413		*Interest on Investments* (gross)		1,332
23		*Post Office Savings Bank Interest*		1
		Institute of Metals:		
1,100		Rent, Rates and Cleaning	842	
625		Library	1,000	
1,725				1,842
—		*Rents Receivable*		775
43		*Sundry Receipts*		32
		Publishing:		
44,495		1956 *Journal*—Revenue	51,634	
8,862		Other Publishing Activities—Revenue	8,373	
—		Adjustments—Amounts over-provided in previous year	22	
53,357				60,029
322		*Translations Service*—Sales		316
£79,982				£87,646
2,738		Balance brought down		1,606
—		Refund of Superannuation Contributions less payments to former members of staff		358
382		Surplus on Scunthorpe Meeting		—
4,904		Excess of Expenditure over Income for the Year		—
£8,024				£1,964

ACCUMULATED FUND, 31ST DECEMBER, 1956

£		£	£		£
2,411	Joint Metallurgical Societies Meeting in Europe	—	36,180	Balance at 31st December, 1955	30,115
250	Transfer to Reserve for Comprehensive Index	250		Joint Metallurgical Societies Meeting in Europe—Net Amount Recovered	157
1,000	,, ,, House Fund	1,000	—	Premium on Conversion of Defence Bonds	10
500	,, ,, Reserve for Supplementing Staff Pensions	500	—	Transfer from Repairs Reserve	—
4,904	Excess of Expenditure over Income for the Year	—	3,000	Excess of Income over Expenditure for the Year	407
30,115	Balance carried to Balance Sheet	28,939	—		
£39,180		£30,689	£39,180		£30,689

HOUSE FUND—COST OF LEASE ACCOUNT

	£
Lease at Cost including Legal Expenses	43,518
	£43,518

	To 31st Dec., 1955 £	Current Year £	Total £
Amount written off:			
Transfer from Accumulated Fund	3,000	1,000	4,000
Bank Deposit Interest	121	—	121
Interest on Investments	928	616	1,544
			5,665
Balance as per Balance Sheet			37,853
			£43,518

ANDREW CARNEGIE RESEARCH FUND

BALANCE SHEET, 31ST DECEMBER, 1956

1955 £	1955 £	LIABILITIES	£	£
	21,241	*Amount of Original Fund* $100,000	21,241	
	4,596	*Add:* Amount capitalized per last Balance Sheet	4,596	
25,837				25,837
		Sundry Creditors:		
	1,682	Grants	1,440	
	16	Auditors' Fees	16	
	133	Iron & Steel Institute—Current Account	—	
1,831				1,456
		Reserve for Travelling Scholarships:		
2,600		Balance at 31st December, 1955		2,600
		Accumulated Fund:		
7,322		Balance per account annexed		7,214
£37,590				£37,107

1955 £	1955 £	ASSETS	£	£
	36,741	*Investments at Cost at 31st December*, 1955	36,741	
		Less Proceeds of Redemption of Defence Bonds	1,995	
		(* Market Value 31st December, 1956—£31,190)		34,746*
		Balances at Bank:		
	188	On Current Account	2,161	
	10	On Deposit Account	10	
	634	Post Office Savings Bank	142	
	832			2,313
	17	*Sundry Debtor*—Income Tax Receivable		34
	—	*Iron and Steel Institute*		14
	£37,590			£37,107

ACCUMULATED FUND, 31ST DECEMBER, 1956

£	£		£	£
2,310		Grants	1,450	
150		Contribution to General Fund on account of Salaries	150	
16		Audit Fee	16	
5		Sundry Expenses	—	
7,322		Balance carried forward	7,214	
£9,803			£8,830	

£	£		£	£
8,340		Balance at 31st December, 1955		7,322
		Income for the year ended 31st December, 1956:		
	1,420	Interest on Investments (gross)	1,483	
	16	Interest on Deposits	8	
1,436				1,491
—		Premiums on Conversion and Profit on Redemption of Defence Bonds		17
27		Premiums on Conversion of Defence Bonds and Serial Funding Stocks		—
£9,803				£8,830

WILLIAMS PRIZE FUND

BALANCE SHEET, 31ST DECEMBER, 1956

1955 £	LIABILITIES	£
	Capital Value representing Market Value of £3,000 3½% Conversion Loan on	
2,220	21st September, 1926 when the fund was inaugurated	2,220
662	Accumulated Fund—Balance per account annexed	587
£2,882		£2,807

1955 £	1955 £	ASSETS	£	£
2,769		*Investments at Cost per Schedule*		2,769
		(*Market Value at 31st December, 1956—£2,535)		
		Balance at Banks:		
	5	On Current Account	38	
	107	Post Office Savings Bank Account	—	
	112			38
		Sundry Debtor:		
	1	Iron and Steel Institute		—
	£2,882			£2,807

ACCUMULATED FUND, 31ST DECEMBER, 1956

£		£	£
300	Awards	200	
662	Balance carried forward	587	
£962		£787	

£	£		£	£
834		Balance 31st December, 1955		662
		Income for year ended 31st December, 1956:		
	121	Interest on 3½% Conversion Loan	121	
	3	Interest on Defence Bonds	3	
	4	Interest on Post Office Savings Bank Account	1	
	128			125
	£962			£787

BESSEMER MEDAL FUND

BALANCE SHEET, 31ST DECEMBER, 1956

1955	LIABILITIES	£	£	1955	ASSETS	£	£
£				£			
	Capital Value representing Market Value of £400 4% Debenture of the L. & N.W. Railway	400			*Investments:* £474 15*s.* 0*d.* British Transport 3% Guaranteed Stock 1978/88	400	
900	*Grant* from The Iron and Steel Institute	500		900	£500 4% Funding Stock 1960/90	500	
			900		(Market Value 31st December, 1956—£718)		900
26	*Accumulated Fund*—Balance per account annexed		34		*Sundry Debtor:*		
				26	Iron and Steel Institute		34
£926			£934	£926			£934

ACCUMULATED FUND, 31ST DECEMBER, 1956

£		£	£	£		£	£
33	Medal	26	42		Balance 31st December, 1955		26
17	Sundry Expenses				*Income for year ended 31st December, 1956:*		
26	Balance carried forward	34		14	Interest on British Transport Stock	14	
				20	,, ,, 4% Funding Stock	20	
			34				34
£76		£60	£76				£60

SIR ROBERT HADFIELD MEDAL FUND

BALANCE SHEET, 31ST DECEMBER, 1956

1955	LIABILITIES	£	1955	ASSETS	£
£			£		
268	*Capital Value* representing Market Value of £267 11*s.* 2*d.* 3% Savings Bonds 1955/65	268	268	£267 11*s.* 2*d.* 3% *Savings Bonds* 1955/65	268
35	Accumulated Fund—Balance per Account annexed	39		(Market Value at 31st December, 1956—£233)	
				Sundry Debtor:	
			35	The Iron and Steel Institute	39
£303		£307	£303		£307

ACCUMULATED FUND, 31ST DECEMBER, 1956

£		£	£		£
3	Medals	4	30	Balance 31st December, 1955	39
35	Balance carried forward	39		Income for year ended 31st December, 1956:	
			8	Interest on 3% Savings Bonds	8
£38		£43	£38		£43

THE IRON AND STEEL INSTITUTE

SCHEDULE OF INVESTMENTS AT 31st DECEMBER, 1956
SHOWING NOMINAL VALUES, COST, AND PRESENT MARKET VALUES

Nominal Value	Nature of Security	Market Value, 31st December, 1956	Cost Value
GENERAL FUNDS OF THE INSTITUTE			
£ s. d.		£	£
2,197 7 0	3½% War Stock	1,550	2,161
1,324 7 4	3½% Conversion Loan	914	1,254
2,954 1 0	4% Consolidated Stock	2,260	3,176
10,042 19 3	4% Funding Loan 1960/90	8,023	7,483
2,000 0 0	3% Savings Bonds 1955/65	1,750	2,000
5,000 0 0	do. 1955/65	4,375	4,972
2,000 0 0	do. 1960/70	1,570	2,000
1,500 0 0	do. 1965/75	1,117	1,500
4,854 2 0	British Transport 3% Guaranteed Stock 1978/88	3,325	5,407
		£24,884	£29,953
LIFE COMPOSITION FUND			
587 13 10	3½% Conversion Loan	415	513
205 15 6	3½% War Stock	145	212
681 13 0	4% Funding Loan 1960/90	535	600
196 10 1	do. 1960/90	155	231
2,615 11 2	British Transport 3% Guaranteed Stock 1978/88	1,792	1,893
		£3,042	£3,449
ANDREW CARNEGIE RESEARCH FUND			
(Trustees: Sir William Larke, K.B.E., The Hon. R. G. Lyttelton, and N. H. Rollason)			
6,897 14 11	3½% War Stock	4,863	6,896
800 0 0	do.	564	794
2,693 12 0	3½% Conversion Loan	1,900	2,007
3,922 14 10	4% Conversion Stock 1957/58	3,883	3,998
1,793 15 0	British Electricity 3% Guaranteed Stock 1968/73	1,355	1,762
4,166 18 7	4% Funding Stock 1960/90	3,272	4,676
500 0 0	4½% Defence Bonds (P.O. Register)	500	500
17,812 10 0	British Transport 3% Guaranteed Stock 1978/88	12,202	10,863
1,000 0 0	3% Savings Bonds 1955/65	875	1,000
2,250 0 0	do. 1960/70	1,766	2,250
		£31,180	£34,746
WILLIAMS PRIZE FUND			
(Trustees: Sir William Larke, K.B.E., The Hon. R. G. Lyttelton, and N. H. Rollason)			
3,452 15 7	3½% Conversion Loan	2,435	2,670
100 0 0	3½% Defence Bonds (P.O. Register)	100	99
		£2,535	£2,769
SIR ROBERT HADFIELD MEDAL FUND			
267 11 2	3% Savings Bonds 1955/65	£233	£268
BESSEMER MEDAL FUND			
(Trustees: Desmond Lysaght, The Hon. R. G. Lyttelton, and N. H. Rollason)			
474 15 0	British Transport 3% Guaranteed Stock 1978/88	325	—
500 0 0	4% Funding Stock 1960/90	393	500
		£718	00
HOUSE FUND			
10,000 0 0	4% Funding Stock 1960/90	7,850	10,131
3,200 0 0	do. 1960/90	2,532	3,251
2,200 0 0	do. 1960/90	1,747	2,030
		£12,129	£15,412

(*Signed*) W. BARR,
Hon. Treasurer

(*Signed*) K. HEADLAM-MORLEY,
Secretary

We have examined the foregoing Balance Sheets and Income and Expenditure Accounts with the Books and Vouchers of the Institute and certify them to be correct. We have also verified the balances at the Bankers and the Securities for the Investments shown above.

FINSBURY CIRCUS HOUSE,
BLOMFIELD STREET, E.C.2
29th March, 1957

(*Signed*) W. B. KEEN & CO.,
Chartered Accountants

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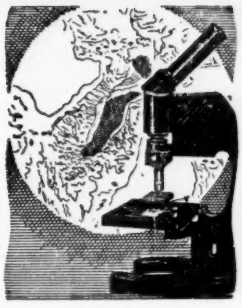


Microhardness Penetration Curves in Two Nickel-bearing Carburized Steels

By Henry M. Otte

SYNOPSIS

The hardness through the case of two carburized steels has been measured after air cooling or water quenching small sections and this has been related to the amount of retained austenite, A, as determined in an earlier investigation, and the carbon content c. The results could be summarized by an empirical equation for the hardness H of the form $H = H_0 + kc - f(cA)$ where H_0 and k are constants. Measurements were made with a diamond pyramid microhardness tester, and the curves obtained are shown to be equivalent to those measured on a taper-ground surface using a standard hardness tester. 1298

IN A CARBURIZED CASE the presence of retained austenite becomes increasingly more prominent as the carbon content is increased above about 0·7%,[1] and this reduces the hardness of the steel since the austenite is relatively soft.[2] Maximum hardness is obtained in a steel by complete transformation to martensite.[3] In most carburized cases the carbon content at the surface is in the region of 1·1%, and consequently appreciable amounts of austenite are usually retained on cooling to room temperature. Large quantities of retained austenite are undesirable because they diminish the beneficial value of carburizing as a means of producing a hard case surrounding a relatively soft core.

Various heat-treatments are available for the reduction or removal of this austenite, but here the topic will be restricted only to observations concerning the influence of the retained austenite (and the carbon concentration) on the hardness of rapidly cooled steels. As studies were made on carburized steels, ranges of carbon and retained austenite contents were examined simultaneously. The steels used and many of the experimental procedures have been described in an earlier publication[1] and will only be briefly referred to here.

EXPERIMENTAL PROCEDURE

Materials and Methods

The chemical composition of the two steels used is given in Table I. Bars (1-in. square cross-section) from each steel were pack-carburized at 900° C in a commercial compound and subsequently sliced into $\frac{3}{16}$-in. sections which were austenitized in a salt bath consisting principally of barium chloride, resistance heated by immersing two graphite electrodes. The carbon penetration curves were determined by chemical analysis of successive layers 0·005–0·010 in. thick, cut from the original bar and the retained austenite by lineal analysis on the sections (*see* ref. 1 for details).

During carburizing, carbides formed rapidly at the surface in the initial stages, but subsequently (for carburizing times in excess of 8 h) their depth of penetration varied linearly with time. The depth to which carbides were visible was fairly uniform and at a magnification of × 500 it could be estimated to within ± 0·002 in. After 120 h carburizing, free carbides could be detected to a depth of 0·025 in. whereas a depth of 0·008 in. had already been reached after 8 h. By austenitizing at a higher temperature (above 900° C) these carbides could be dissolved, provided that the maximum solid solubility in the austenite was not exceeded. When the carbides were present in the microstructure together with retained austenite and martensite, very erratic hardness readings were generally obtained. The present investigation was therefore carried out with the carbides in solution. This had the further advantage that the exact composition of the austenite was thereby known.

The variation of hardness with depth was measured with a microhardness tester across a section of the carburized steel. In making such measurements attention had to be given to several points to obtain reliable values, and these are described in greater detail under the next heading.

Microhardness Measurements

These were made with a Cooke, Troughton and Simms microhardness tester as described by Taylor[4]

Manuscript received 20th March, 1956.

Dr. Otte is I.C.I. Research Fellow at the Department of Physical Metallurgy, University of Birmingham.

Table I

CHEMICAL ANALYSIS, %, OF ALLOYS STUDIED

Steel	C	Si	Mn	S	P	Cr	Ni	Mo
K (Krupp)	0·12	0·22	0·40	0·018	0·017	1·24	3·80	0·07
N (Nickel)	0·18	0·21	0·41	0·013	0·014	0·32	4·89	0·29

mounted on a Vickers projection microscope using a standard 136° diamond pyramid indenter (set in the objective holder) under a direct load of 200 g. The size of the indentation was measured at a magnification of × 1000 ± 5 to the nearest 0·5 mm on a millimetre grid screen. Careful examination of the diamond indenter revealed no apparent faults such as loose seating, bad point, or poor finish. External vibrations caused a marked lowering of the hardness and were eliminated by placing the whole Vickers projection microscope on a shock-absorbent rubber cushion.

To operate the microhardness tester accurately and thus obtain reliable and reproducible readings, the following precautions were necessary:

(i) Firm seating of the specimen in the instrument (failure to observe this resulted in ' double ' impressions and/or impressions that were badly distorted and frequently too large)

(ii) Careful balancing of the instrument before application of the load (a sensitivity of 0·4 g or less, could easily be obtained)

(iii) Alignment of centre of gravity of applied load with axis of indenter

(iv) Slow application of the load (to avoid inertia effects)

(v) Application of full load (this was ensured by lowering the specimen 5μ past the point at which contact with the indenter was established, as indicated by the extinction of a small electric light. Lowering by more than 30μ past this point affected the readings measurably).

(vi) Standardization of indentation period to about 10 s (longer periods had no effect unless creep occurred, which in the present material was not observed for indentation periods up to 1 min)

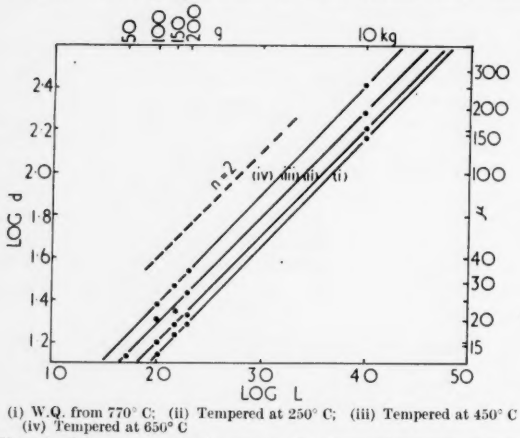
(vii) The use of a sufficiently heavy load to minimize the effect of friction which causes sticking and prevents full penetration.[5]

With these precautions it was found that the standard deviation of 10 readings was 7% or less. This meant that an individual reading would have a 95% probability of being ± 7% within the average.

Since controversy existed[6] regarding the variation of hardness with the load applied to a pyramid indenter, this variation was briefly re-examined by making measurements on a 0·9% plain carbon steel, W.Q. and tempered to various temperatures.

It was found that the Vickers hardness number (VHN) was not affected by load, and Fig. 1 shows that a log–log plot of load *v.* indentation diagonal yielded the theoretical slope of $n = 2$ (for the Meyer index) over the whole range of values up to and including a reading obtained with a standard Vickers machine employing a 10-kg load. This is in agreement with the observations of other investigators.[2, 7] Micro and macro VHN's were therefore directly comparable.

All indentations were made on a mechanically polished surface. If any appreciable amounts (15% or more) of retained austenite were present, this produced a thin layer of slightly cold-worked material at the surface and consequently higher VHN's at low loads due to incomplete penetration of the cold-worked layer. In a fully austenitic specimen this was sufficient to reduce the value of n from 2 to about 1·85–1·90. With 20–25% retained austenite the value of n was reduced to about 1·94.



(i) W.Q. from 770° C; (ii) Tempered at 250° C; (iii) Tempered at 450° C (iv) Tempered at 650° C

Fig. 1—Meyer analysis on 0·9% homogeneous plain carbon steel after heat-treatment as shown

RESULTS

Comparison of the Micro- and Macrohardness Penetration Curves

The usual hardness/penetration curve is most frequently obtained from measurements on a taper-ground surface.[8, 9] In this method the indentations (usually fairly large) made into the case are in the same direction as that in which the carbon diffused during carburizing and hence also in the direction in which the hardness is changing. The alternative method (employed here) is to make micro-indentations

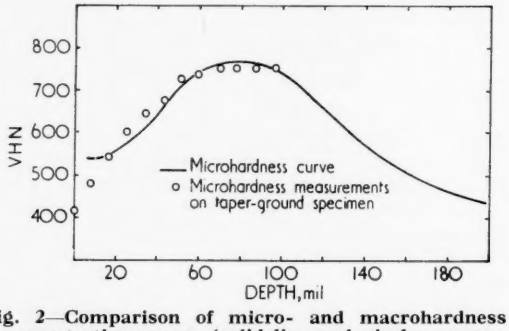
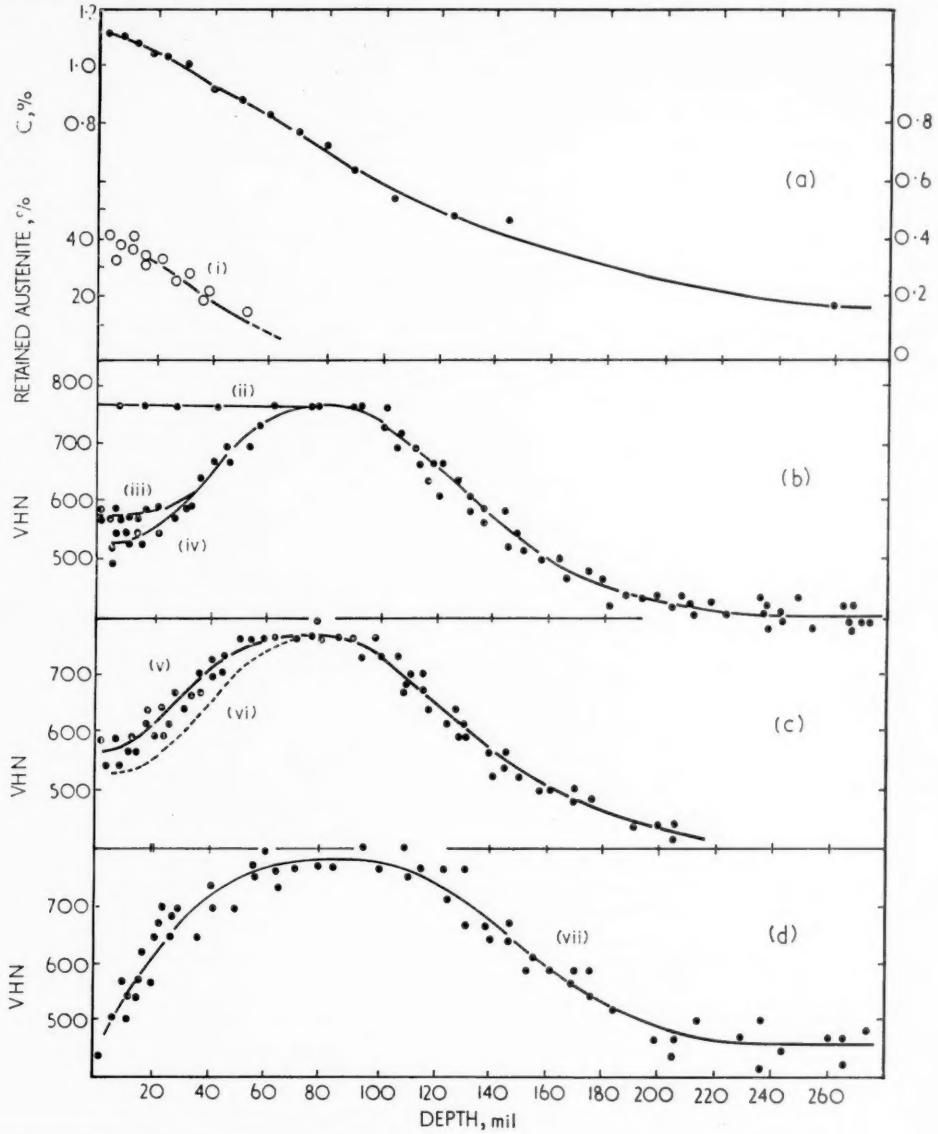


Fig. 2—Comparison of micro- and macrohardness penetration curves (solid line and circles respectively) in carburized steel *N*, A.C. after 5 min at 1000° C

on a cross-section, i.e. essentially into material of uniform hardness in the direction of indentation. Difficulty in interpretation is immediately apparent if the hardness curve obtained by indenting along the tapered surface does not agree with that obtained by indentation across a section. The equivalence of the two methods in the present case was therefore established in the following way. A $\frac{3}{16}$-in. thick slice from steel *N* was air-cooled after 5 min at 1000° C, then one of the side faces ($\frac{3}{16}$ × 1 in.) was taper-ground to a gradient of 1 in 10. A series of hardness readings was obtained along this face using a standard Vickers testing machine and a 10-kg load (these are the macrohardness measurements shown as circles in Fig. 2). From the adjacent face, representing the cross-section of the bar, microhardness readings were obtained. The resulting curve, Fig. 2, coincides fairly well with the macrohardness readings, although agreement would probably not have been so good if breakdown of the case had



	Austenitizing temp., °C	Time, s	Method of cooling
(i)	1100	2	A.C.
(ii)	790	10	A.C.
(iii)	1100	2	A.C.
(iv)	900	5	A.C.
(v)	900	30	A.C.
(vi)	900	5 and 15	A.C.
(vii)	900	5	W.Q.

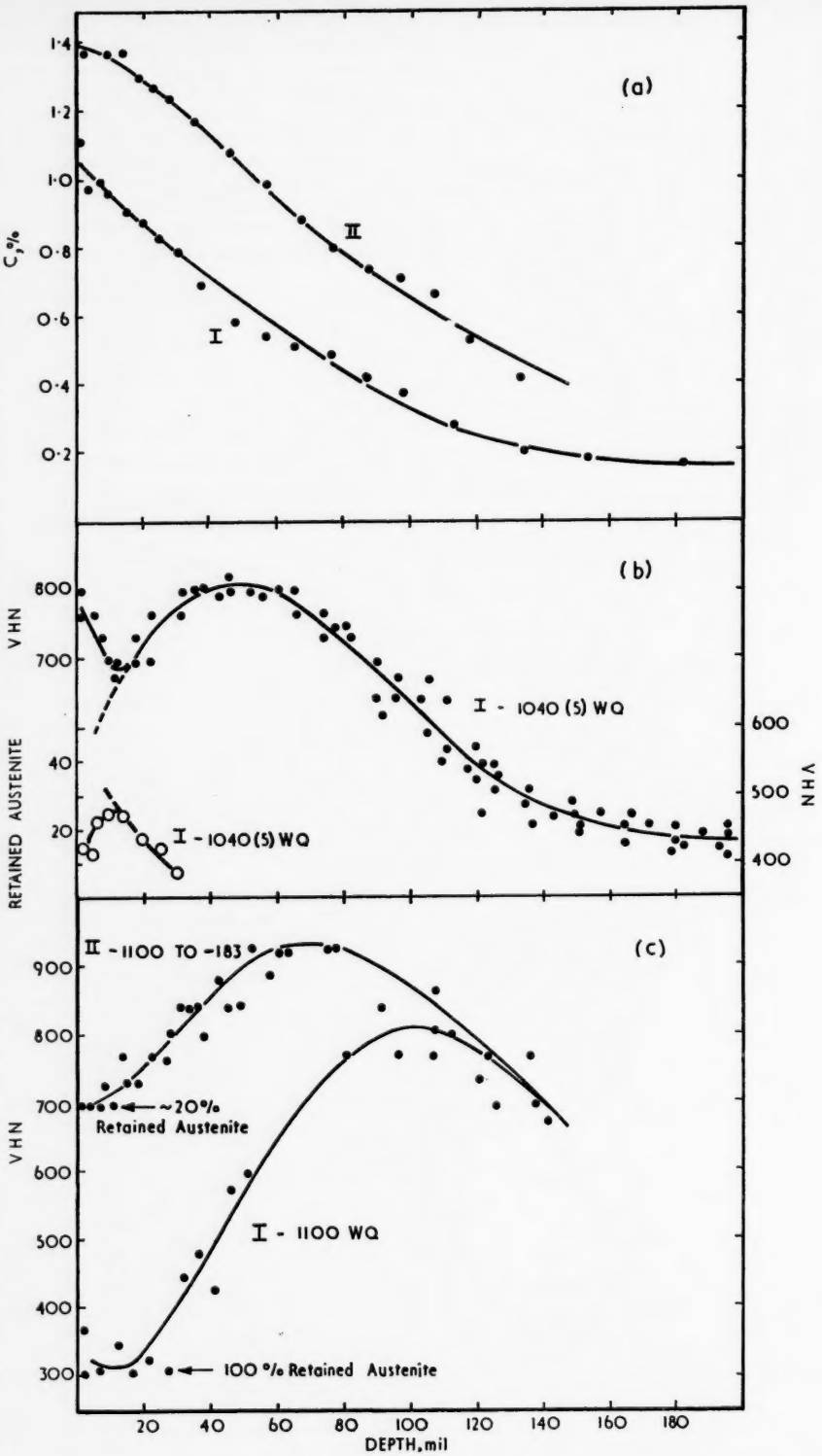
Fig. 3—Carbon, retained austenite, and hardness *v.* penetration curves in steel *N*

occurred. This, however, was unlikely as Kühnert[10] had found that with a 14-kg load breakdown occurred only if the case was less than 0·016 in. (0·4 mm) thick. Furthermore, on doubling the case depth the breakdown load increased five-fold. As the macrohardness curve shown in Fig. 2 was obtained with a 10-kg load on a carburized case more than 0·160 in. thick, no possibility existed of the case collapsing. Discrepancy between the two methods of determining the hardness/penetration curve may arise if appropriate care is not taken in the preparation of the specimens. Thus if a section has to be cut from the carburized bar after the final heat-treatment has been completed, temper ing may occur during the sectioning. It is therefore not surprising to note that Pomp and Kirsch[11] obtained a lower hardness by indenting across a section than by indenting into the case at successive levels. The method of indenting across a section and thus into material of uniform hardness (as used in the present work) is to be preferred over the alternative method because of its greater sensitivity to small hardness variations such as may result for example from slight decarburization at the surface, as shown in Fig. 4*b*.*

Hardness/Penetration Curves

For a given carbon gradient it was found that the general shape of the hardness/penetration curve was reproducible

* A different salt bath (containing a patent mixture of chlorides) was employed to austenitize this specimen.



(i) Carburized 90 h in a compound with reduced activity
(ii) Carburized 120 h at 900° C then treated 1½ h at 1100° C in pure charcoal

Fig. 4—Carbon, retained austenite, and hardness *v.* penetration curves in steel *K*

within fairly narrow limits. The important variables were austenitizing temperature and time, cooling rate, and quenching temperature. These will be considered briefly in turn below and illustrated with a few selected results. The specimens had the carbon penetration curves indicated in Figs. 3*a* and 4*a*.

Under the experimental conditions employed, no change in the hardness/penetration curve was observed for times up to 15 min at 900° C. After 30 min at 900° C, a small change resulting from carbon diffusion and consequent change in the carbon penetration curve was detected, but only between the surface and a depth corresponding approximately to the maximum hardness which was found to occur at about 0·7% carbon, as shown in Fig. 3*c*. A comparable observation was made by Rowland and Lyle[12] who noted no change in case depth at the 0·5% carbon level in a carburized Krupp steel after 45 min at 900° C.

At 1100° C no difference in the hardness curves could be detected between austenitizing times of 40 and 60 s.[1] However, a 2-min austenitization at 1000° C seemed to produce a small amount of diffusion outward. The amount of retained austenite at the 1% carbon level (about 0·015 in. from the surface) dropped from 44 to 35% as a result and the hardness increased from 540 VHN to about 580 VHN,

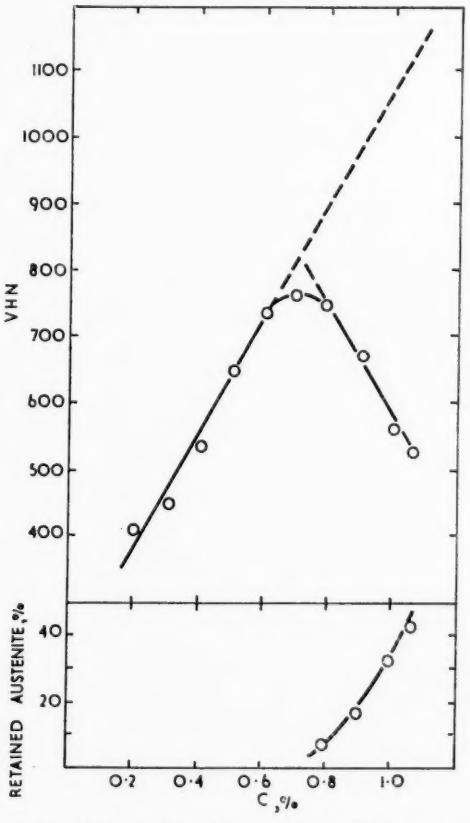


Fig. 5—Variation of hardness and retained austenite with carbon, steel *N*, A.C. from 900° C

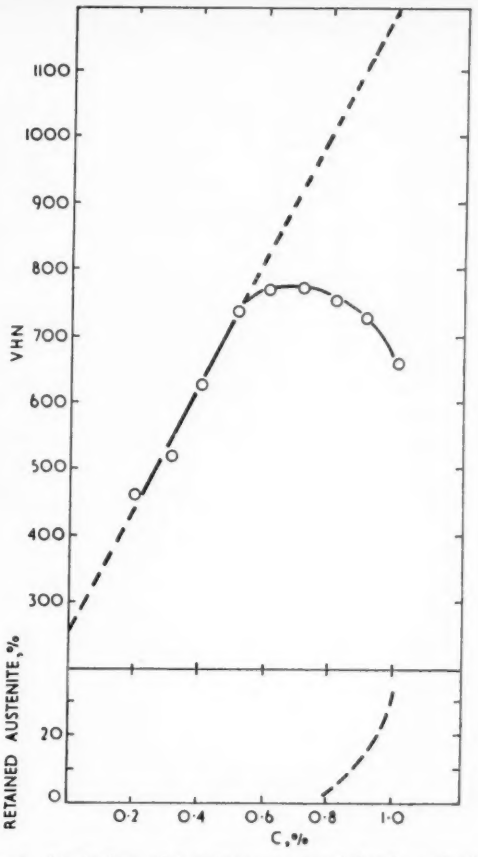


Fig. 6—Variation of hardness and retained austenite with carbon, steel *N*, W.Q. from 900° C

but at a depth below the 0·9% carbon level (about 0·040 in. from the surface) no change was observed. Figure 3*a* shows the complete retained austenite gradient in the case.

The diffusion of carbon outwards is essentially a form of decarburization and has been reported to occur in apparently neutral atmospheres.[13, 14] The formation of a graphite film on a 1·15% plain C steel heated into the γ region and cooled in high vacuum has also been observed[15] and considered characteristic of high-carbon steels and not due to external or vapour-phase contamination.

Increasing the austenitizing temperature increases the diffusion rate of carbon and thus brings about the expected more rapid destruction of the carbon gradient established by carburizing. Quenching from different temperatures in the fully austenitic region is known to have no effect on the amount of retained austenite at room temperature.[1] Thus, for a given austenitizing time the highest convenient temperature is to be preferred (for the present purpose) since the rapid solution of any undissolved carbides is then most favoured. If the austenitizing times are kept short (2–5 min) the carbon gradient as reflected in the hardness measurements can be essentially preserved. Figure 3*b* shows this to be true except at the very

Table II
VALUES OF CONSTANTS IN EQUATION (1)*

Steel	Treatment	H_0	k
K	W.Q.	260	1040
N	W.Q.	260	940
N	A.C.	210	870

* Constants obtained from Figs. 5–7

surface, where noticeable diffusion had occurred at 1100° C, as already mentioned above.

From Fig. 3*b* it may also be seen that austenitizing at 790° C yielded a case of uniform hardness to quite a large depth. Virtually no retained austenite was present; the structure consisted essentially of martensite and finely dispersed carbides, to be expected when austenitizing below a temperature at which all carbides are in solution.

Water-quenching instead of air-cooling from 1100°C produced a hardness curve with essentially the same peak hardness, but with a more gradual drop in hardness on either side of it. In effect this meant that the faster cooling rate produced a higher hardness throughout the whole case. Examples of microhardness penetration curves of W.Q. and A.C. specimens are reproduced in Fig. 3.

If any retained austenite was present at room temperature, cooling further to liquid-air temperature gave an increase in hardness that would be expected to result from the transformation of austenite to martensite. In a specimen from a bar carburized 120 h at 900° C, 100% austenite was retained at the surface on water quenching from 1100° C.* The

* Pure powdered charcoal was used in place of the salt bath for this experiment.

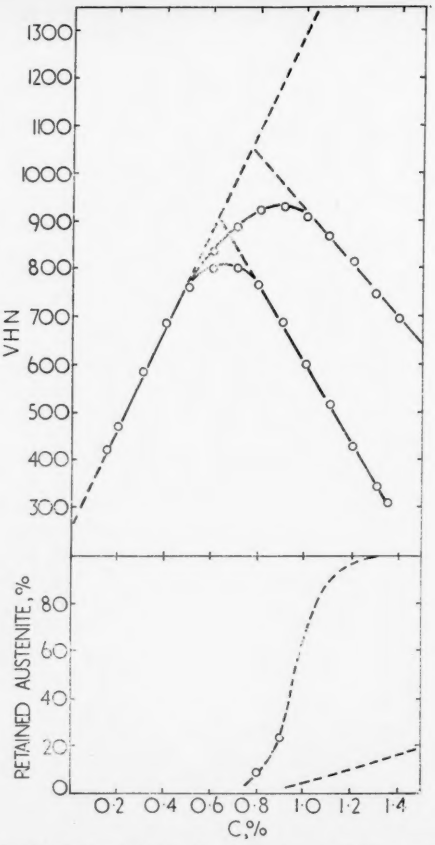
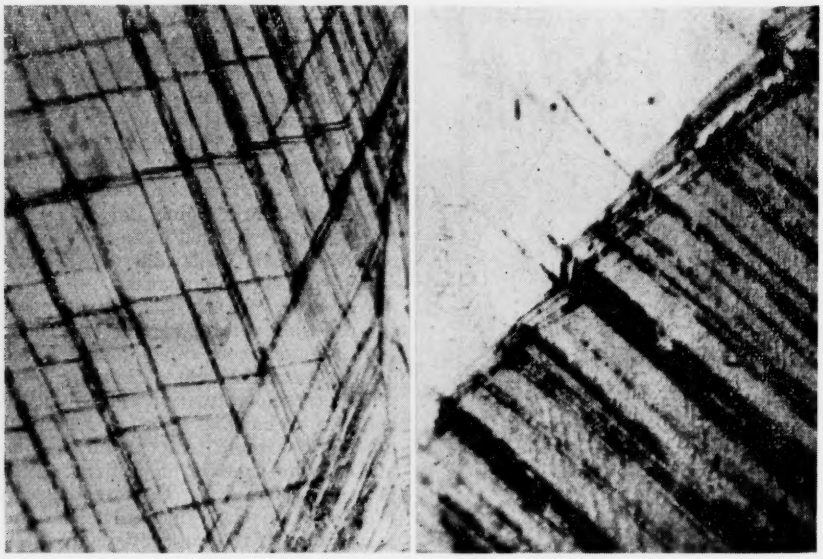


Fig. 7—Variation of hardness and retained austenite with carbon, steel *K*, W.Q. from 1040–1100° C



(*a*) Displacement of one set of deformation bands as produced by second set (*b*) Stepped grain boundary produced by deformation band

Fig. 8—Deformed region around indentation in austenite of a 1·5% C, 5% Ni steel, W.Q. from 1100 °C, etched with 1% nital after tempering 45 s at 300° C and electropolishing × 350

Table III

REDUCTION (H_A) IN HARDNESS OF KRUPP STEEL DUE TO RETAINED AUSTENITE

Retained austenite, %	H_A*	
	Room temperature	— 183° C
5	280 (0·78)	375 (1·00)
10	380 (0·82)	660 (1·18)
20	485 (0·89)	1170 (1·50)

* Carbon content, %, in parentheses

Table IV

VALUES OF CONSTANTS IN EQUATION (4)*

Steel	Treatment	n	R	Approximate range of validity, % retained austenite
N	A.C.	$\frac{3}{4}$	35	0–40
N	W.Q.	$\frac{1}{3}$	170	0–30
K	W.Q.	$\frac{1}{3}$	180	0–80
K	—183° C	$\frac{3}{4}$	120	0–20

* Constants obtained from Figs. 5–7

hardness (300 VHN) at the surface was raised to 700 VHN by the transformation of 80% of the retained austenite after immersion in liquid air, as shown in Fig. 4*c*. In addition, the peak hardness was increased and shifted closer to the surface, reflecting the effect of carbon in increasing the hardness of the martensite.

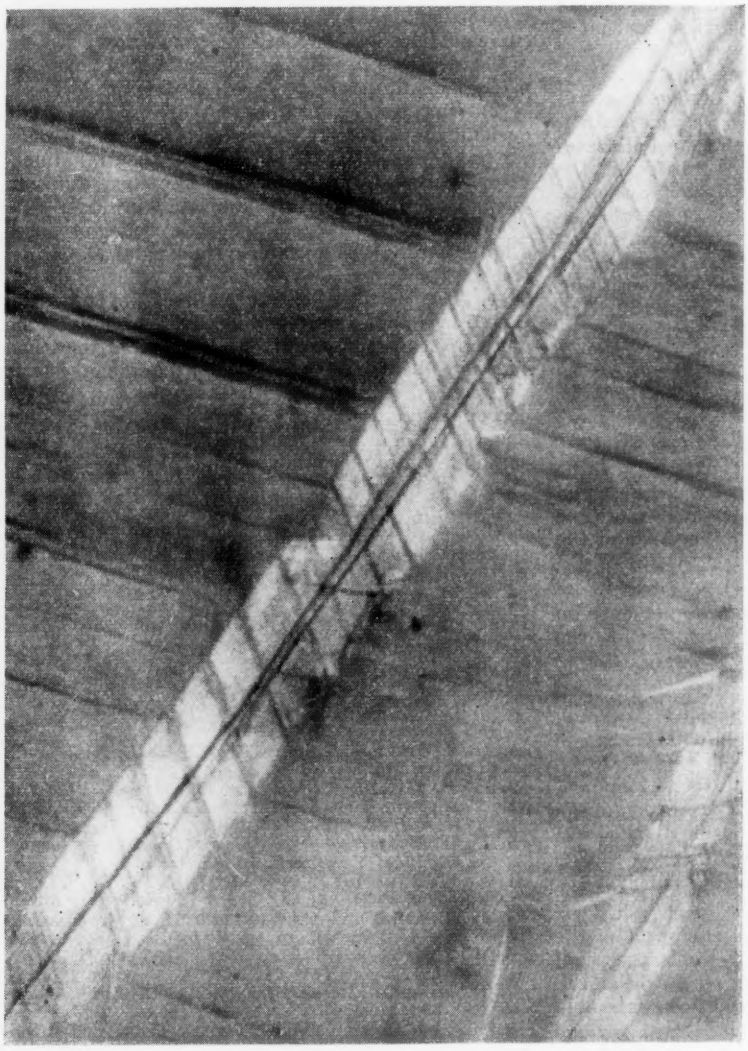


Fig. 9—Deformed martensite crystal in a 1·5% C,5% Ni steel, W.Q. from 1100° C, untempered. Photograph shows split mid-rib and what are probably thin twins running across the plate, × 1200, enlarged 3 diameters

Variation of Hardness with Carbon Content

This is shown in Figs. 5–7 for both the steels after water quenching, and for steel *N* after air cooling. The data for the curves were obtained from Figs. 3 and 4. The retained austenite curves shown in Figs. 6 and 7 have been taken from published data.[1]

A maximum hardness was generally observed at about 0·7% carbon after air-cooling or water-quenching the specimens. At this carbon content the amount of retained austenite was too small to be estimated reliably under the microscope.

Below 0·7% carbon, a martensitic structure was obtained, the hardness of which decreased with decreasing carbon content. But for a given carbon content in this range, water quenching produced a slightly higher hardness than air cooling, probably due to tempering of the martensite which occurs during air cooling.* The small amount of retained austenite which has been shown to exist below 0·7% carbon[16] may, however, play quite as important a part as tempering in affecting the hardness. The effect of tempering is expected to be smaller when the carbon content exceeds 0·7%, because of the lowering of M_s.

Above 0·7% carbon, a reduction in hardness occurs due to retained austenite, the

* Also at very low carbon contents, probably due to a small amount of intermediate transformation of the austenite.

amount of which increases with increasing carbon content. Again, for a given carbon content, water quenching showed a higher hardness than air cooling, but this time it was due to the lesser amount of austenite retained on water quenching. This is a reflection of the lesser amount of stabilization, and is in agreement with previous observations.[1]

DISCUSSION

Effect of Carbon on Hardness of Martensite

Figures 5–7 show that for a fully martensitic structure, the relationship between hardness (H_M) of the martensite and its carbon content c is about linear, as is apparent also from published work (e.g. refs. 3, 17), and can thus be represented by a function of the form

$$H_M = H_0 + k\,c \quad \ldots\ldots\ldots\ldots\ldots\ldots (1)$$

where H_0 is the hardness extrapolated to zero carbon content, and k is a constant. Table II shows that if martensite could be obtained in a carbon-free iron, its hardness would probably lie around 260 VHN. Equation (1) is valid up to 0·5% carbon and still fairly accurate up to 0·6%. Above this carbon content the actual hardness of the steel deviates markedly from the extrapolated hardness of the martensite. This is largely due to the presence of retained austenite in the steel, but may also be partly the result of a decreasing effect of carbon on the hardness of the martensite. Since the austenite cannot be completely transformed to martensite by cooling to − 183° C, the hardness of a fully martensitic steel in this high-carbon range cannot be measured. Individual martensite plates are too small to give a reliable hardness value, though attempts to measure it have been made[2, 18]

Effect of Retained Austenite on Hardness of Steel

Assuming the extrapolation of equation (1) to be valid, a quantity H_A representing the reduction in hardness due to the retained austenite may be evaluated. Thus if H is the actual hardness of the steel,

$$H_A = H_M - H \quad \ldots\ldots\ldots\ldots\ldots\ldots (2)$$

Table III shows that H_A is dependent not only on the amount of retained austenite, but also on the carbon content of the retained austenite, i.e.

$$H_A = f(c\,A) \quad \ldots\ldots\ldots\ldots\ldots\ldots (3)$$

where A is the percentage of retained austenite. For a given amount of retained austenite, H_A increases with increasing carbon content.

If for the moment the effect of carbon is ignored, then H_A can be represented by an expression of the form

$$f(A) = RA^n \quad \ldots\ldots\ldots\ldots\ldots\ldots (4)$$

where n and R are constants. Table IV gives the values of n and R obtained from the curves in Figs. 5–7. It is to be noted that since the effect of the carbon content has been ignored, the constants n and R of the W.Q. steel K are altered on refrigeration. The quantitative effect of the carbon content could not be reliably evaluated because of the small quantity of austenite retained after the sub-zero quench.

The actual values of n and R are probably of very little significance; what they emphasize, however, is that as the amount of retained austenite is decreased its effect on lowering the hardness becomes more pronounced. A physical picture of the reason for this may be obtained from the microscopic inspection of the deformation produced by an indentation. This reveals that the austenite flows plastically as shown in Fig. 8, producing strain bands analogous to those found in α-brass,[19] while within the martensite thin twins appear to form (Fig. 9) as was also reported by Lucas.[20] The removal of the last few per cent of austenite thus probably also removes any 'lubricating' effect that the austenite may have exerted in allowing the martensite to deform readily.

CONCLUSIONS

(1) If the case of a carburized steel is sufficiently thick (so that breakdown of the case does not occur when indenting into it) then the hardness penetration curve obtained by microhardness tests across a section of the case is essentially the same as that obtained by macro-indentations into the case at different levels.

(2) Hardness/penetration curves indicate that austenitizing ($\frac{3}{16}$-in. thick sections) for short periods (up to 30 min at 900° C) in essentially neutral atmospheres causes a small amount of carbon diffusion outwards, but no measurable amount inwards.

(3) For specimens $\frac{3}{16}$ in. thick, air cooling produces a lower hardness than water quenching at all carbon levels.

(4) The decrease in the hardness of the steel caused by retained austenite depends not only on the amount of austenite retained but also on its carbon content where this exceeds 0·7% carbon. Thus the removal of 80% retained austenite at the 1·4% carbon level resulted in a proportionately smaller increase in hardness than did 20% at the 0·9 carbon level (in a Krupp-type steel).

Acknowledgments

This work formed part of a research programme sponsored by British Timken Ltd. and carried out in the Physical Metallurgy Department of Birmingham University, under the general direction of the late Professor D. Hanson. The author wishes to thank Mr. D. McNicoll and Mr. J. H. Evans for works facilities and to other members of the staff for their co-operation.

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20. F. F. LUCAS: *Trans. Amer. Soc. Steel Treating*, 1924, vol. 6, pp. 669–686 (Fig. 23).

Effect of Alloying Elements on the High-temperature Tensile Strength of Normalized Low-carbon Steel

By J. Glen, B.Sc., A.R.T.C., F.I.M.

SYNOPSIS

The object of the investigation was to study the effect of individual alloying elements on the high-temperature tensile strength of normalized low-carbon steel. For this purpose a series of steels containing varying amounts of alloying elements was prepared. Some of these steels were made both with and without the addition of aluminium. The elements investigated were manganese, chromium, molybdenum, tungsten, vanadium, titanium, silicon, nickel, and copper. It was found that for a simple iron–carbon alloy one strain-age-hardening effect was obtained with a maximum at about 200° C. With a little nitrogen present this strain-age-hardening effect was intensified. With the addition of either manganese, chromium, molybdenum, tungsten, or copper, an additional strain-age-hardening effect was obtained at a temperature higher than 200° C, the actual temperature depending on the element used. A strain-age-hardening effect due to vanadium or titanium was not obtained with the experimental conditions used. It was also found that the alloying elements modified the strain-age-hardening due to carbon and nitrogen.

Corresponding to each strain-age-hardening effect at a certain temperature a minimum in reduction of area was also obtained. It was concluded that the maxima in stress and the minima in ductility were related to some form of coherent precipitation of carbides or alloy carbides in the dislocations. 1332

STEEL, in common with most metals, is used primarily because of its strength and its ability to deform and work-harden without fracture. Although these properties have been studied ever since steel was first made, it is only some 50 years since it was realized that metallic crystals deform by a process of slip on crystallographic planes. The fundamental mechanism of the process has proved more difficult to resolve, but it is now generally accepted that the explanation lies in the behaviour of lattice defects which have been called dislocations. Even so, some problems such as work-hardening are very complex and still defy solution. To quote Cottrell,[1] on work-hardening "it was the first problem to be attempted by the dislocation theory of slip, and may well prove the last to be solved."

The effect of chemical composition and heat treatment on the mechanical properties can be studied in many different ways. Of these, the tensile test has always been one of the most common, and its importance can be measured by the fact that it is mandatory in practically all materials' specifications. The information obtained has proved very useful in design; for example, the working stress of most structures is based on a factor of safety of four on the tensile strength. This means that under perfect conditions each individual part of a structure is stressed well below its elastic limit.

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The author is with the Research Department of Colvilles, Ltd.

In most engineering structures, however, the material is fabricated to shape, so that some degree of work-hardening may result. Also, since it is virtually impossible to eliminate all points of stress concentration, localized plastic deformation takes place when the structure is loaded. Localized plastic deformation also occurs in the vicinity of welds. To this might be added a novel type of work-hardening

now being encountered in materials used for the construction of atomic power plants. This hardening results from the bombardment of the material by nuclear particles thrown off during irradiation processes.[2]

From the above it is clear that work-hardening and related phenomena, such as strain-age-hardening, are of importance in practice and must be considered in studying the effect of alloying elements and heat treatment on the mechanical properties. This is all the more important to-day because of the rapid advance in engineering technology and the necessity of utilizing each steel to its greatest advantage by increasing the working stress to the maximum compatible with safety.

Reference to the literature shows that the effect of alloying elements on the conventional tensile strength has been studied in some detail, but much less attention has been given to their effect on strain-age-hardening. Edwards and others,[3, 4] using tensile tests at ambient temperature, concluded that alloying elements which do not form carbides tend to increase strain-ageing, but their effect is generally small. Carbide-forming elements decrease the amount of strain-age-hardening, and the more powerful carbide formers, such as titanium, completely suppress the ageing, and at the same time eliminate the yield-point phenomenon. Edwards concluded that strain-ageing resulted from the precipitation of carbides and nitrides.

More recently, Cottrell[1] has suggested a different mechanism for these phenomena. He postulates that interstitial solute atoms such as carbon and nitrogen tend to form atmospheres round the dislocations present in the steel. The stress required to move a dislocation with an atmosphere is greater than that required once the dislocation has moved away from its atmosphere, so that in a tensile test at room temperature an upper and lower yield point can be obtained. When a test piece is strained, say 5%, the yield point disappears, but given time the solute atoms migrate back into the dislocations so that the yield-point phenomenon returns and the material strain-age-hardens. In a test carried out at about 200° C, the rate of diffusion of carbon and nitrogen is such that the atmospheres can migrate as fast as the dislocations, so that strain-age-hardening is instantaneous. The main point in favour of this theory is the fact that it explains how strain-ageing can be obtained even when the amount of solute present is less than the solubility limit. Although the Cottrell theory seems to explain many of the facts of room-temperature ageing, it has not yet been expanded to include the effect of alloying elements.

Using true-stress/true-strain tensile tests, the author investigated the influence of manganese and molybdenum on the high-temperature tensile properties of low-carbon steel.[5] It was shown that, in the absence of manganese and molybdenum, a maximum in the stress for a given strain was obtained at about 200° C, and at the same time a minimum was obtained in the ductility. When manganese was added to the steel, the above maximum in stress persisted, but a second strain-age-hardening phenomenon appeared, having a maximum effect at about 300° C. When molybdenum was added instead of manganese the second maximum occurred at about 500° C. Since these secondary strain-ageing phenomena resulted from the presence of alloying elements, it was surmised that precipitation of alloy carbide occurred during the process of straining.

This work has now been extended to include the effect of most of the common alloying elements used in steel. The effect of tempering and cold working on both room-temperature and high-temperature properties has also been investigated. A large number of creep tests have also been carried out. The present paper, however, is limited to a consideration of the effect of individual alloying elements on the high-temperature tensile properties of low-carbon steel in the normalized condition. The following alloy elements have been investigated: manganese, chromium, molybdenum, tungsten, vanadium, titanium, silicon, nickel, and copper.

Test Procedure

As soon as yielding commences in a tensile test, the cross-sectional area of the test piece begins to diminish. Thus, with increasing deformation, the load divided by the original cross-section of the test piece becomes an increasingly inaccurate measure of the stress on the material. This does not greatly detract from the usefulness of the conventional tensile test, since for similar materials the elongation at the maximum load does not vary greatly. However, when comparing steels of widely different compositions and treatment, the elongation at the maximum load may vary from as little as 1% to as much as 50%. In such cases a comparison of the tensile strength may be quite misleading. For the present work, therefore, it was considered necessary to use a more accurate measure of the stress/strain relationship, so that the true stress could be compared at various known levels of strain.

All the tests were carried out on a 20-ton Avery multilever tensile testing machine using 0·564 in. diameter test pieces, finished by grinding to obtain exactly 3 in. parallel between the shoulders. For the room-temperature tests a Lindley extensometer was attached to the gauge length of the specimens, and, in addition, two dial gauges were connected to the crosshead of the tensile machine, one reading to 0·001 in. and the other to 0·0001 in. The crosshead straining rate was 0·015 in./min. Readings on the Lindley extensometer were ended at about 1% extension, but readings of the dial gauges were taken at frequent intervals throughout the test. The test was stopped at the yield point and again at frequent intervals, so that simultaneous readings of the load and the diameter of the test piece could be measured.

The true stress/strain curve was calculated by the method proposed by MacGregor[6] i.e. using $\ln Ao/A$ or $\ln L/Lo$, where Ao and Lo are the original area and gauge length respectively, and A and L the area and length at any instant. Up to the maximum load, the gauge length remains essentially parallel, so that either area or length may be used to calculate the strain. This has been verified for many of the tests carried out. From the maximum load to fracture, area alone can be used, since, due to necking, the change in length is no longer a measure of the true strain. As was pointed out by MacGregor[6] and confirmed by all the room-temperature tests carried

out for this paper, the stress/strain curve from the maximum load to fracture is very nearly a straight line. Advantage was taken of this fact to determine the true stress/strain curve at high temperatures From the length measurements it was possible to define the stress/strain curve as far as the maximum load, and from the load and area at fracture it was possible to define the breaking point. A straight line from maximum load to fracture then gives a good approximation to the true stress/strain curve in this region. From the room-temperature tests and from a few checks at high temperature, where the test was discontinued at an intermediate point, it was found that, when the reduction of area was about 50% or less, both methods gave very nearly the same results. With a greater reduction of area, the stress at fracture estimated from the breaking load and final area tends to be overestimated. The error was found to be approximately 3 tons/in^2 at 70% reduction of area. For very high reduction of area the accuracy of measurement of the fracture stress was poor. However, in most of these cases the stress at fracture was less than that at the maximum load, because of recrystallization of the test piece.

In testing most of the steels, particularly at about 200° C, pronounced jogging of the load occurred, due to strain-age-hardening effects. In all such cases a smooth curve was drawn through the plotted points of load against extension and the true stress/strain curve estimated from this average curve.

In all the high-temperature tests a standard procedure was used to exclude as far as possible any extraneous effects. For example, variations in stress could result from variations in the rate of straining and variations in the time at temperature before testing. The procedure used was as follows. The furnace was heated to approximately the correct temperature and the test assembly inserted. The specimen reached the desired temperature within approximately half an hour and was maintained at temperature for half an hour before commencing the test. The furnace was thermostatically controlled to ± 2° C. The crosshead straining speed was maintained constant at a rate of 0·015 in./min. Obviously the rate of straining of the test piece will vary from this value, particularly after the maximum load, when straining occurs only in the portion which is being necked down. Creep will also occur to some extent in the tests, depending on the temperature and the composition of the steel. For this reason the strength values obtained at high temperature are only correct for the conditions of testing used. It is believed, however, that this does not in any way influence the conclusions obtained.

Presentation of Data

In a previous paper by the author on the effect of manganese and molybdenum, the data were summarized in the form of iso-strain curves, since the main object of that paper was to demonstrate the existence of secondary strain-age-hardening phenomena. In the present more detailed paper, however, it was considered advisable to show the actual true stress/strain curves. In the body of the paper, only sufficient curves have been included to illustrate the variations. For clarity of presentation individual tests have been displaced along the axis at a distance proportional to the temperature of testing. The temperature of testing is indicated on the curves.

At high temperature, the proof stress (0·2% extension or less) appears to vary quite markedly owing to minor metallurgical effects. This aspect of the matter is being made the subject of a separate study and will not be considered in this paper, which deals with higher strain values. However, the lower yield point (if present) was clearly indicated by the testing procedure and has thus been included on the graphs. The points on all the temperature stress/strain curves correspond to 0·5%, 1%, 2%, 3%, etc. extension, the maximum load being indicated by the last point shown. The fracture point is also indicated. The points on the room-temperature curves, however, do not refer to any particular strain, since they were obtained from measurements made of load and diameter at intervals throughout the test.

From the graphs of the true stress/strain tests, the stresses corresponding to the lower yield point, 0·01 strain (1% extension), 0·1 strain (10·5% extension), and 0·5 strain (39·2% reduction of area) were measured, and graphs of stress against temperature plotted to show the iso-strain curves at these levels of strain. It should be noted that in nearly all cases the 0·01 and 0·1 strain values were clearly defined by the elongation measurements. The 0·5 strain values are less accurate since the values are influenced by any error in measurement of the fracture stress or strain.

A further comparison of the effect of individual elements was made by plotting the true strain at fracture (i.e. the reduction of area) against the testing temperature.

Details of Experimental Steels

The tests originally carried out were exploratory in character, since very little was known about the high-temperature properties of steel. Much of the test material used was collected over a long period of time, being stock remaining from experimental and production casts made either in a 6-cwt high-frequency furnace or a 10-ton basic electric-arc furnace. Some tests were also carried out on production casts made in an 80-ton basic open-hearth furnace. As the results accumulated it seemed worth while making a more detailed study of the subject. For this purpose a number of 25-kVA high-frequency furnace casts were prepared and tested to augment the data which had already been obtained.

The small high-frequency casts were obtained in the form of 18 or 27-lb ingots which were forged to 1-in. square bar. All the other material was obtained in the form of 1-in. to 1$\frac{1}{8}$-in. dia. rolled bar.

The carbon content of all the steels is approximately 0·1%. This restriction arose originally because the author was primarily interested in steels for use at high temperature, which are usually of low carbon content. Such steels, however, have the advantage that appreciable quantities of alloying elements can be added without unduly hardening the steel.

The sulphur contents of the steels range from 0·026% to 0·042%, and the phosphorus from 0·014% to 0·038%. The content of residual elements in most of the steels was very low (less than 0·01% of

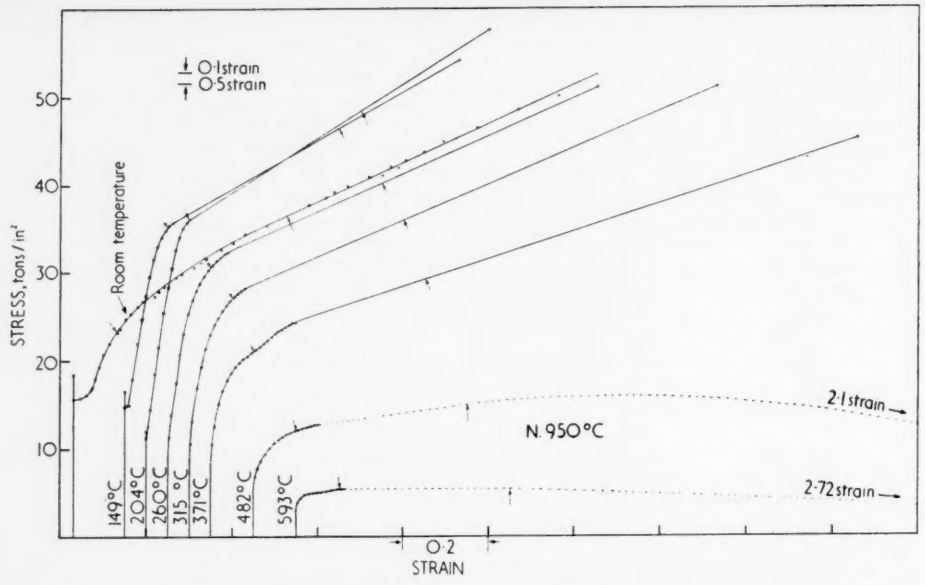

Fig. 1—True stress/strain curves for steel 1 (0·11% C, 0·07% Mn)

any element) since most of the high-frequency casts had been prepared using pure Armco iron scrap. In a few of the more recently prepared steels, the residual alloy content is somewhat higher, but since this appears to have had little or no effect on the results, the analysis of each steel is not given in detail.

All the steels made to coarse-grained practice have been numbered *1*, *2*, *3*, etc. The titanium steels, which of necessity were deoxidized with about 2 lb/ton of aluminium, are included in this sequence. Because of the importance of aluminium additions, a smaller series of steels was prepared, where a deliberate addition of 3 lb of aluminium per ton was made. All these steels showed a McQuaid–Ehn grain size of 8. Such steels are numbered *1A*, *2A*, etc. For convenience, the carbon, silicon, manganese and alloying element content of each steel is given later, when discussing the particular alloying element concerned.

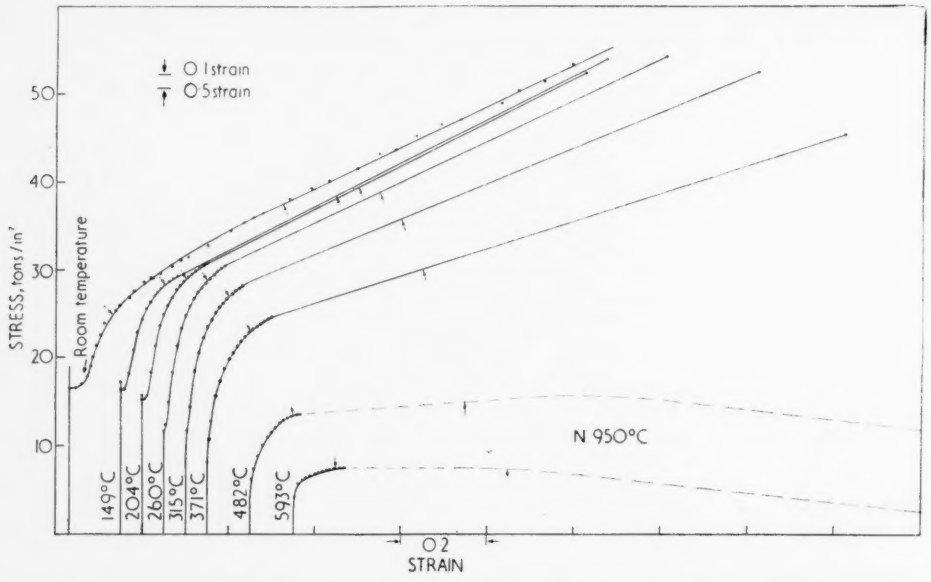

Fig. 2—True stress/strain curves for steel 1A (0·105% C, 0·15% Mn) deoxidized with aluminium

Heat Treatment

As already indicated, the present paper is limited to a consideration of steels tested in the normalized condition. Two exceptions have been made to this rule for reasons which will be apparent. It was recognized that the normalizing temperature used might influence the results obtained, but in the first instance it was decided that only one normalizing temperature would be used for any individual steel. In most cases the normalizing temperature was 950° C. This temperature was not high enough to cause undue coarsening of the steel, but at the same time it ensured that a uniform austenite would be obtained with all the carbides in solution. For the vanadium steels a higher temperature (1000° or 1050° C) seemed necessary. In the case of titanium it was considered more interesting to determine the properties of the steel in a condition where any carbon and nitrogen out of solution was likely to be in the form of titanium carbide and nitride. By normalizing at 950° C a close approximation to this condition should have been obtained. The lowest titanium steel was, however, also heat-treated at 1150° C to compare the effect of titanium with the other alloying elements.

To carry out the heat treatment, the bars were cut into 10-in. lengths and placed in the electric furnace, which was controlled thermostatically. The appropriate normalizing temperature was maintained for a period of 1 h before the specimens were removed and allowed to cool freely in air.

LOW-MANGANESE STEEL

Two steels of low manganese content were tested to obtain a datum line with which the results on alloy steels could be compared. Details of these steels are:

Steel	Type	C, %	Si, %	Mn, %	Al Addition, lb/ton
1	H.F.	0·11	0·07	0·07	0·5
1A	H.F.	0·105	0·10	0·15	3·0

The true stress/strain curves for these steels are shown in Figs. 1 and 2, and the derived iso-strain curves (0·01, 0·1, and 0·5 strain) in Fig. 3. The variation of reduction of area with temperature is shown in Fig. 4.

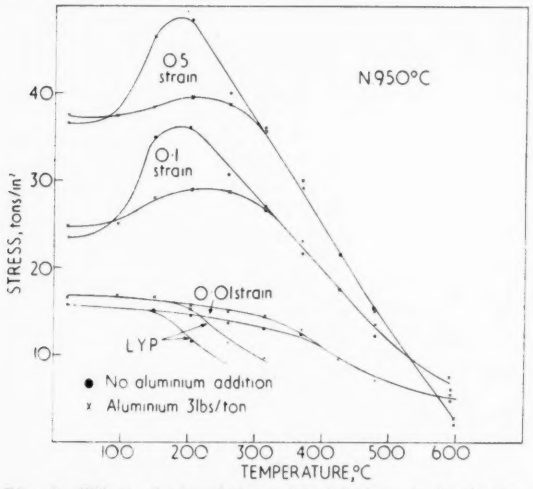

Fig. 3—Effect of aluminium deoxidation on 0·1% Mn steels (1 and 1A)

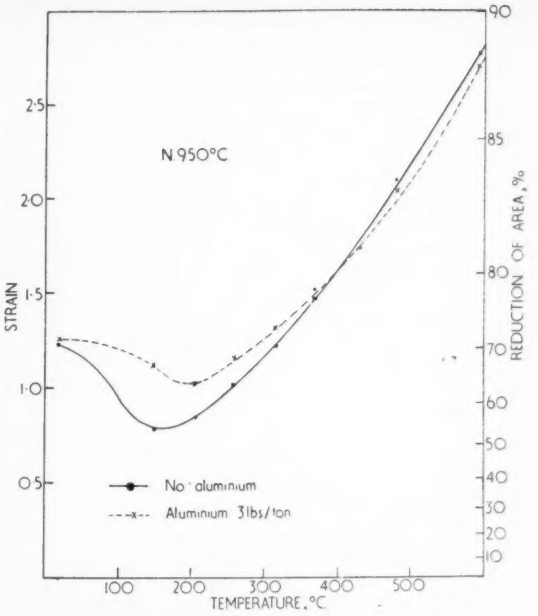

Fig. 4—Ductility of 0·1% Mn steels (1 and 1A)

It will be noted from Fig. 1 (steel *1*) that a large yield-point elongation is obtained in the test at room temperature. This elongation decreases with increasing temperature of testing and eventually disappears. At temperatures somewhat above room temperature the rate of work-hardening in the initial stages of testing is increased, compared with that for the room-temperature test. In these tests the rate of work-hardening decreases more or less abruptly at the maximum load, and from that point until fracture occurs is not much different from that of the room-temperature test. For the aluminium-killed steel *1A* (Fig. 2) similar remarks apply, but in this case the initially fast rate of work-hardening in the high-temperature tests is not maintained, so that although the stress/strain curves for strains up to about 0·05 are little different from those of steel *1*, at higher strains steel *1* is stronger. In both steels, at temperatures above about 250° C, the stress for a given strain decreases more or less rapidly with increasing temperature of testing.

From Fig. 3 it will be seen that at 0·01 strain there is little to choose between the steels, although it should be noted that the yield-point phenomenon persists to a higher temperature with steel *1A*. At 0·1 strain a maximum in stress is obtained for both steels at about 200° C. With steel *1A*, however, the maximum is less pronounced and appears to occur at a slightly higher temperature compared with steel *1*.

At 0·5 strain similar results are obtained. Above about 250° C there is no difference between the steels as judged from the results on this graph.

The reduction in area (Fig. 4) shows a minimum at about the same temperature as the maximum in stress.

This minimum is less pronounced in the case of steel *1A* and appears to occur at a slightly higher temperature. Above about 300° C there is little difference between the steels, as was also found to be the case for the stress values.

As already indicated, dislocation theory has yet to explain many of the complexities of the work-hardening process. However, for the present purpose it suffices to accept as proven that a pure metal, e.g. iron, contains a large number of dislocations which can move under the influence of a small applied stress. Thus pure iron has a low elastic limit. As deformation proceeds, multiplication of dislocations occurs and many become locked because of mutual interference, so that the material work-hardens. With increasing temperature of testing, there is a gradual decrease in the stress for a given strain, i.e. there are no strain-age-hardening phenomena.

In a previous paper[5] the author suggested that coherent precipitation of carbide and nitride occurs during the process of straining steel at high temperature. Cottrell[1] however considers that atmospheres of carbon or nitrogen atoms form round the dislocations and thus increase the stress required to move them. More recently, Dahl and Lucke[7] and Thomas and Leak[8] have shown that the dislocation atmosphere is too dense to be considered as such, and thus precipitation must occur. Obviously the real nature and form of the atmosphere or precipitate must await a more complete understanding of the basic atomic processes. However, it was considered more correct and simpler for the present purpose, particularly since alloy carbides and nitrides are involved, to use the term ' dislocation precipitate ' rather than ' atmosphere.'

In steel *1A*, much of the nitrogen in the steel is present as aluminium nitride and so little remains in interstitial solution.[9] On testing at room temperature (Fig. 2) a high stress is necessary to move the dislocations away from their associated carbide precipitates, but, once the dislocations are free, a lower stress is sufficient to keep them moving. Considerable deformation occurs at this lower yield point, but is eventually stopped by the normal mechanism of work-hardening. On further straining, the rate of work-hardening is initially high but decreases with increasing deformation, until at about the maximum load the rate of work-hardening becomes constant. (By correcting for the triaxial stresses set up by necking of the test piece the rate of work-hardening decreases until fracture occurs.)[10]

Because of the large number of dislocations created during straining, some of the carbon atoms in solution may precipitate in the dislocations even at room temperature, but the process is not complete. Thus, if a test piece is deformed say 10%, the carbon in solution, if given sufficient time, can migrate and precipitate in the dislocations. The original dislocation precipitates, having lost their associated dislocations, are unstable so that they tend to redissolve, migrate, and reprecipitate in other dislocations.

At a somewhat higher temperature, the rate of diffusion of carbon in steel is increased so that dislocation precipitates can form during testing. As at room temperature, an upper and lower yield point is obtained, but the amount of elongation at the yield point is reduced compared with that at room temperature, since almost instantaneous precipitation in the dislocations augments the normal strengthening effect of work-hardening. The lower yield point eventually disappears because of thermal effects which decrease the stability of the dislocation precipitate.

On straining beyond the yield point, the rate of work-hardening is increased compared with that at room temperature because the carbon in solution precipitates in the dislocations during straining. Two other factors must be considered. First, the solubility of carbon is increased compared with room temperature so that more carbon is available for precipitation. Secondly, it seems conclusive from the work of Cottrell and Leak[11] on quench-ageing and strain-ageing that dislocation precipitates (in their paper they use the term ' atmospheres ') are more stable than fine precipitates not associated with dislocations. Thus there is a tendency for solution of any excess carbide which then reprecipitates in the dislocations.

The high rate of work-hardening on straining at high temperature is maintained as long as carbon is available for precipitation. The supply is limited, however, so that after a certain amount of straining most of the carbon is used up in the dislocations which have been created. On further straining, still more dislocations are formed which are not impeded by precipitate, so that the rate of work-hardening decreases to a value comparable with that obtained for the same strain at room temperature, where little or no ageing occurs during testing. The maximum degree of strain-age-hardening occurs at a certain temperature which will vary with the rate of straining. The lower this rate, the lower the temperature at which the maximum effect occurs, since it is controlled by the rate of diffusion of carbon. At still higher temperatures, more carbon is available for precipitation, but thermal effects have also increased, so that the dislocation precipitates become increasingly less stable. Due to continual resolution, reprecipitation, and spheroidization, the dislocations are hindered to a lesser extent at high temperatures of testing. Thus, with increasing temperature above about 200° C, there is a rapid fall in the stress for strains greater than about 0·1. The decrease in stress with temperature is not marked for very small strains and this can be explained as follows. At low strains the number of dislocations is relatively small and the amount of carbon available is relatively large. Thus, as fast as carbon atoms go into solution from the dislocation precipitates, they are replaced. As the strain increases the number of dislocations multiplies but the carbon does not increase, so that as a precipitate dissolves due to thermal effects there is less chance of replacement. The dislocations can thus move more easily.

In steel *1*, containing nitrogen in solution as well as carbon (Fig. 1), the test at room temperature is very little different from that of steel *1A* (Fig. 2). This is to be expected since there is sufficient carbon alone to saturate the dislocations present in the unstrained state, so the addition of nitrogen has relatively little effect. During straining, interstitial elements have no time to migrate so that little ageing occurs and the stress/strain curve is almost the same as for steel *1A*.

Table I

COMPARISION OF WEIGHT AND NUMBER OF ATOMS, %

Weight %	Atomic %									
	C	Mn	Cr	Mo	W	V	Ti	Si	Ni	Cu
0·1	0·465	0·102	0·1075	0·058	0·0305	0·1095	0·1165	0·200	0·095	0·088
1·0	..	1·027	1·084	0·588	0·307	1·108	0·118	2·010	0·961	0·887
2·0	..	2·074	2·196	1·187	0·619	2·237	2·237	4·061	1·941	1·793

In fact, because of its finer ferritic grain size, the aluminium-killed steel tends to have a slightly higher yield point than steel *1* despite the absence of interstitial nitrogen.

At higher temperatures the yield point persists as with steel *1A* but disappears at a lower temperature because the strain-age-hardening effect of nitrogen is additional to that of carbon in augmenting the ordinary work-hardening effect. Similarly, on straining beyond the yield point, the high rate of work-hardening is maintained to a greater strain because the presence of nitrogen increases the amount of available solute and more dislocations can be filled with precipitate. As before, however, the supply of interstitial solute is limited, and the rate of work-hardening decreases abruptly at about the maximum stress, and thereafter is more comparable to that of the test at room temperature. Above about 200° C the stress for a given strain decreases similarly to steel *1A*. Above 300° C there is little or no difference between the two steels. The rate of diffusion of nitrogen is greater than that of carbon, so that the nitride precipitate is less stable than the carbide. Moreover, only about 0·005% nitrogen is present, and this is all in solution at these temperatures. Thus, above 300° C the strength is mainly due to carbon and there is no difference between the steels.

Since the rate of diffusion of nitrogen is somewhat greater than that of carbon, the maximum in stress occurs at a slightly lower temperature with steel *1* as compared with steel *1A* (Fig. 3). In steel *1*, therefore, there appear to be two superimposed maxima which are not differentiated by the testing technique used. Similarly, the minimum in reduction of area for steel *1* occurs at a slightly lower temperature than that for steel *1A* (Fig. 1). It might be expected that steel 1 would show a minimum in reduction of area since the fracture stress at 200° C is increased considerably, but this does not explain the minimum obtained for steel *1A*, where the fracture stress decreases with increasing temperature. Above about 300° C there is little to choose between the steels, since the controlling factor seems to be thermal effects rather than the presence of interstitial elements.

COMPOSITION OF ALLOY STEEL

In the tables below, the analyses of the steels are given in conventional form. In all cases, however, it should be realized that the amount of alloying elements present when measured by weight per cent may be very different from that obtained when comparison is made on the basis of the number of atoms of each element present. In Table I, the weight per cent is compared with atomic per cent of the various alloying elements used. The atomic per cent was calculated on the basis that the remainder of the alloy is iron, this being sufficiently accurate for the purpose.

As is shown, all of the above alloying elements, apart from carbon, silicon, molybdenum and tungsten, are of much the same atomic weight, so that the weight per cent and atomic per cent are approximately the same. Molybdenum and tungsten have greater atomic weights than the other metals, so that when the number of atoms is important an additional 70% by weight of molybdenum and 230% by weight of tungsten is required to give the same effect as a given percentage of, say, manganese or chromium.

MANGANESE

Six steels were used to investigate the influence of manganese, two of these being deoxidized with aluminium. Details of these steels are shown below, the low-manganese steels being included for comparison:

Steel	Type	C, %	Si, %	Mn, %	Al Addition, lb/ton
1	H.F.	0·11	0·07	0·07	0·5
2	B.O.H.	0·10	0·05	0·43	0·5
3	B.O.H.	0·13	0·05	0·59	0·5
4	H.F *	0·13	0·10	1·50	0
5	H.F.	0·09	0·29	3·70	0
1A	H.F.	0·105	0·10	0·15	3
2A	B.O.H.	0·10	0·06	0·45	3
4A	H.F.	0·105	0·18	1·53	3

* 6 cwt. H.F. Furnace

The results obtained for steels *2*, *3*, and *2A* are not markedly different from those of steels *1* and *1A*. The true stress/strain curves for steels *4* and *4A* are shown in Figs. 5 and 6 and those for steel *5* in Fig. 7. Iso-strain curves for steels *2* and *2A* are shown in Fig. 8 and those for steels *4* and *4A* in Fig. 9. The effect of manganese is summarized by the iso-strain curves in Fig. 10. The variation of reduction in area with temperature is shown in Fig. 11.

The addition of 1·5% manganese to a steel lowers the transformation temperature on normalizing so that the carbide structure is finer. As shown by Gensamer,[12] the finer the dispersion of carbide, the greater the tensile strength. Despite this increased strength the general features of the stress/strain curves for steel *4* (Fig. 5) are similar to those of the low-manganese steel *1* (Fig. 1). A notable difference

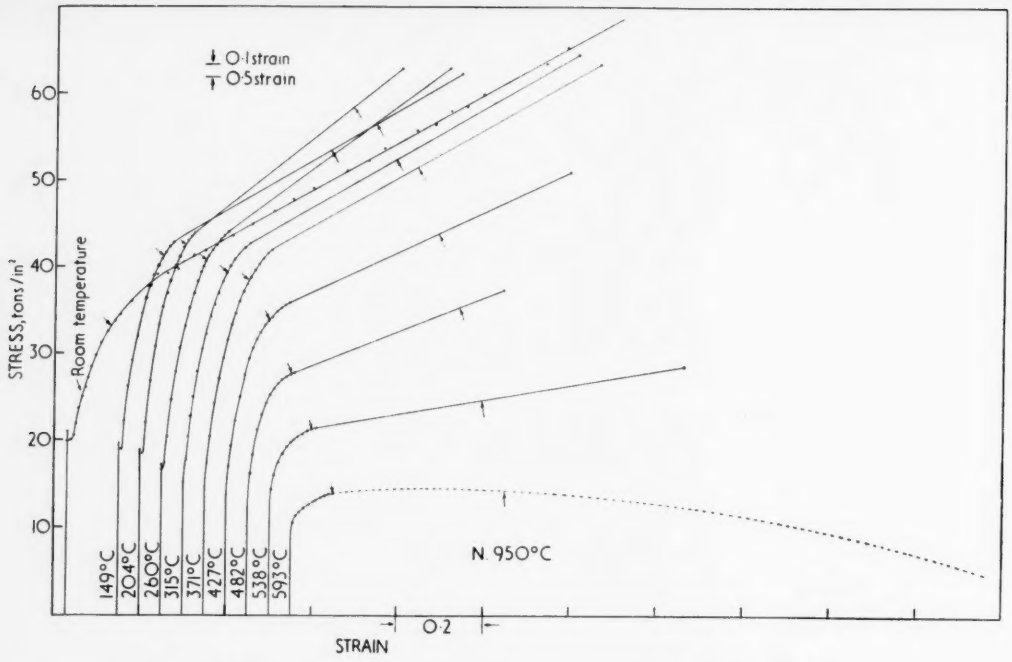

Fig. 5—True stress/strain curves for steel 4 (0·13% C, 1·50% Mn)

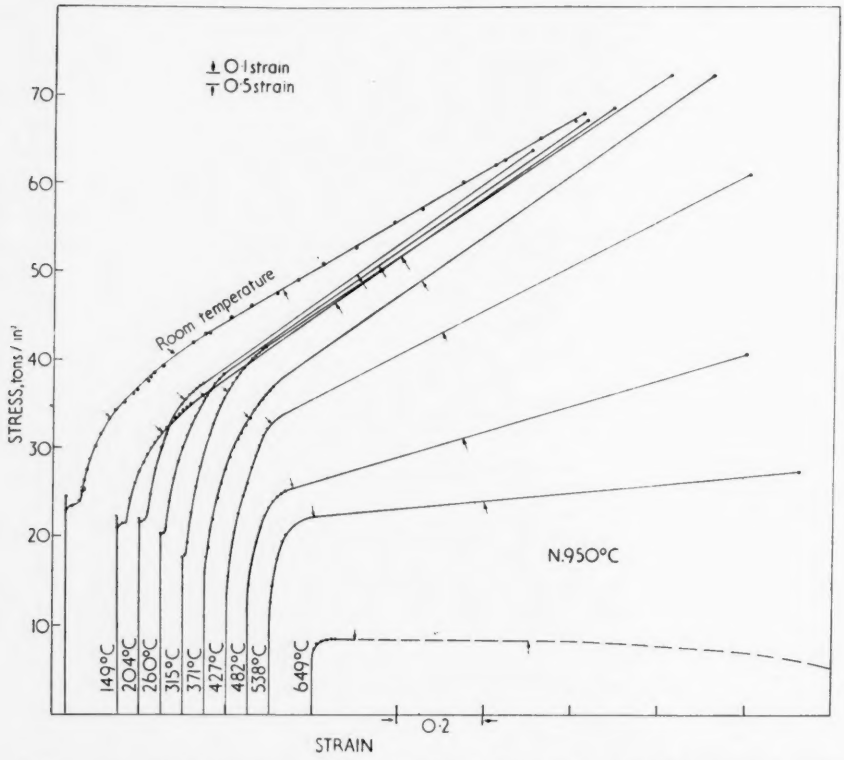

Fig. 6—True stress/strain curves for steel 4A (0·105% C, 1·53% Mn) deoxidized with aluminium

is the fact that the high rate of work-hardening in tests above room temperature persists to a much higher temperature than with the low-manganese steel.

For the aluminium-killed steel *4A* (Fig. 6), the stress/strain curve at room temperature is very similar to that of steel *4* except that the yield point is somewhat higher. As already noted, a similar result was obtained for steels *1* and *1A*. At somewhat higher temperatures steel *4A* behaves in a similar manner to the low-manganese steel *1A* (Fig. 2), but above about 200° C the rate of work-hardening in the initial stages of testing increases markedly compared with that of tests at lower temperatures. In this temperature range the results are similar to those of steel *4*, i.e. the addition of aluminium has little effect on the strength of the steel.

The results for the 3·7% manganese steel (Fig. 7) are notable for the very high strength at small deformation. Above about 0·05 strain the rate of work-hardening at any temperature is not much different from that of the 1·5% manganese steel. The room-temperature test was made on a test piece with shoulders of large radius. With the standard test piece, premature failure took place at the shoulder after less than 5% extension. The fracture had a bright crystalline appearance.

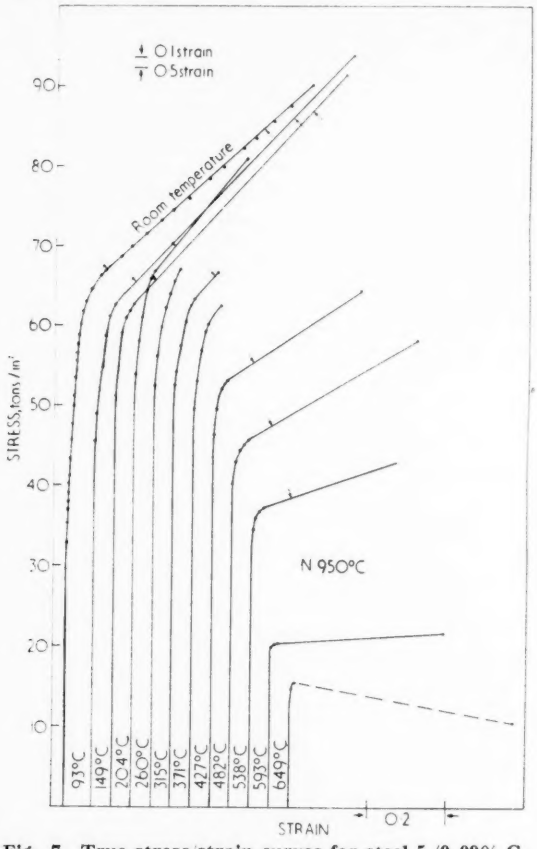

Fig. 7—True stress/strain curves for steel 5 (0·09% C, 3·7% Mn)

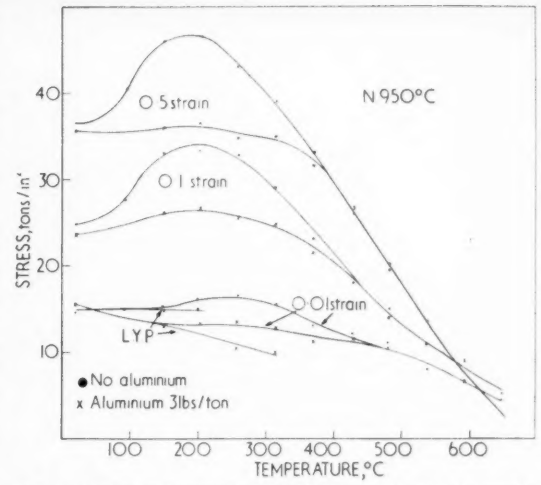

Fig. 8—Effect of aluminium addition on 0·5% Mn steels (2 and 2A)

The iso-strain curves for the 0·45% manganese steels (Fig. 8) are very similar to those of the low-manganese steels (Fig. 3), and confirm that above about 300° C the addition of aluminium has no effect on the strength of the steel. The data for the 1·5% Mn steels (Fig. 9) agree with this conclusion. Figure 9 also shows that manganese increases the strength at high temperature, the effect being a maximum between 300–400° C. The maximum in stress, already noted at about 200° C for the low-manganese steel, does not appear to be greatly influenced by the addition of manganese. In both Figs. 8 and 9 it will be noted that the yield point persists to a higher temperature when the steel is deoxidized with aluminium. This was also true of the low-manganese steels (Fig. 3).

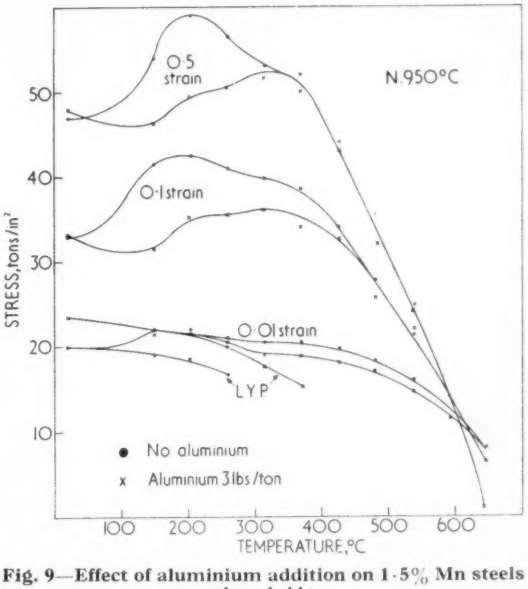

Fig. 9—Effect of aluminium addition on 1·5% Mn steels (4 and 4A)

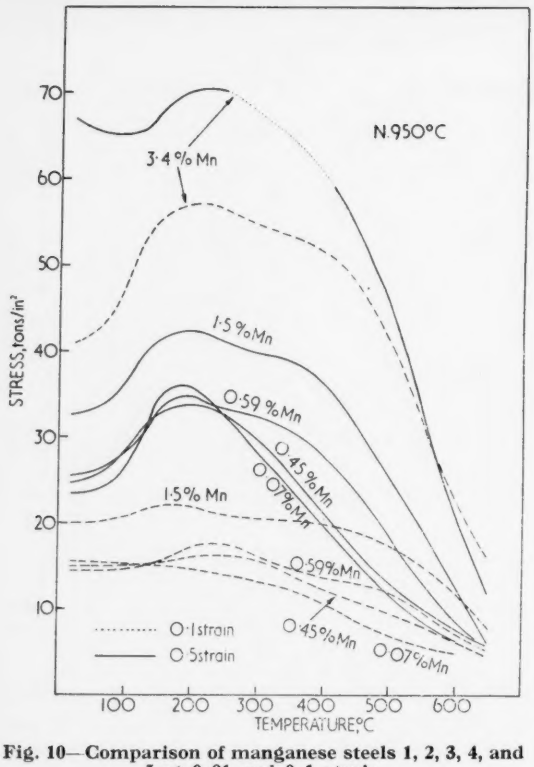

Fig. 10—Comparison of manganese steels 1, 2, 3, 4, and 5 at 0·01 and 0·1 strain

The effect of manganese is more clearly demonstrated by the iso-strain curves (0·01 and 0·1 strain) for steels *1–5* in Fig. 10. At 0·01 strain (1% extension) manganese up to 0·6% has little effect at room temperature, but the higher-manganese steels, particularly the 3·7% manganese steel, are much stronger. At temperatures above about 200° C the 0·1 strain values increase with increasing manganese content. At 0·1 strain a maximum in stress is obtained at about 200° C. The height of these maxima above the stress for 0·1 strain at room temperature decreases with increasing manganese, and the maximum appears to occur at a slightly higher temperature. The strengthening effect of manganese at temperatures above 200° C is evident. Because of the maximum in stress at about 200° C, an actual maximum is not obtained due to manganese. It is obvious, however, that manganese has a maximum effect at about 300–400° C as evidenced by the results on the aluminium-killed 1·5% manganese steel (Fig. 9).

When manganese is added, the curves of reduction in area still show a minimum value at about 200° C (Fig. 11), but a second effect is observed at higher temperatures. For the lower-manganese steels this is simply a point of inflexion in the curve, but with 1·5% and 3·7% manganese steels a second minimum is obtained at about 500° C. As was noted for the maximum in stress, the minimum at about 200° C appears to occur at a higher temperature with increasing manganese content. The addition of aluminium increases the reduction of area, but whereas for the 0·1% manganese steels there is little or no difference in reduction of area above about 300° C, for the manganese steels the difference is appreciable, the aluminium-killed steels being the better. At about 600° C, however, the curves for both the aluminium-killed and ordinary steels tend to the same values.

The factors determining the solubility of an element in interstitial solution in a metal are still largely unknown, but of necessity the interstitial atoms must be small. The stability of the co-existing phase is also important. Thus, although the oxygen atom is smaller than either the nitrogen or the carbon atom, the solubility of oxygen in the range of testing temperature is near zero, because iron and oxygen have such a great affinity that all the oxygen is present as oxide.[13] At the same time there must be some affinity between the interstitial atoms and the solute atoms, otherwise the solubility would again be zero. For example, the helium atom is smaller than the carbon or nitrogen atom but has such a highly stable configuration of two paired electrons that it does not dissolve in any metal.[13]

As shown by Fast and Dijkstra[14] using internal friction measurements, it is more favourable for the nitrogen atoms to occupy an interstice where they have manganese atoms as immediate neighbours (" manganese site ") rather than an interstice which is surrounded exclusively by iron atoms (" iron site "). The nitrogen atoms are able to migrate round the manganese atoms, i.e. into adjacent interstices, but have difficulty in moving away from a " manganese site." At room temperature, therefore, fewer nitrogen atoms are available for strain-age-hardening. A similar argument probably applies to carbon since manganese carbide is more stable than iron carbide. This explains Edwards's conclusion,[4] that the amount of strain-age-hardening decreases with increasing

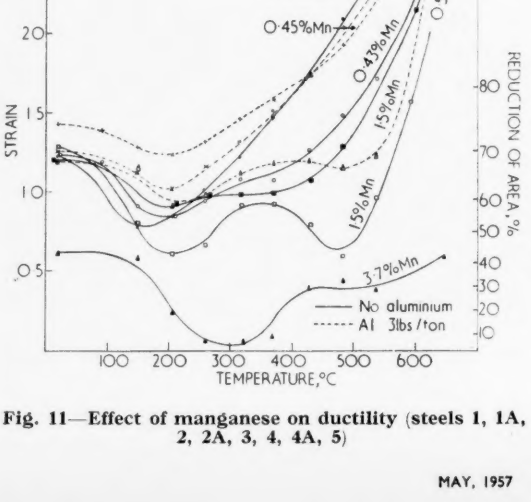

Fig. 11—Effect of manganese on ductility (steels 1, 1A, 2, 2A, 3, 4, 4A, 5)

manganese content. This effect is equivalent to a reduction in the rate of diffusion of carbon and nitrogen, which explains the observation already made that the maximum in stress at about 200° C is reduced and shifted to a slightly higher temperature as the manganese content increases (Fig. 10). The minimum in reduction of area at about 200° C also shifts to a higher temperature with increasing manganese for the same reason.

As already stated, above about 200° C the iron carbide dislocation precipitates become increasingly unstable and less effective in preventing deformation (Fig. 3). When manganese is added, however, another phenomenon must be considered. On normalizing a steel containing manganese, the rate of cooling is too rapid to permit the formation of the equilibrium amount of manganese carbide. On tempering, manganese atoms can diffuse, and there is a gradual solution of iron carbide and precipitation of the more stable manganese carbide.[15] No doubt a similar argument applies to nitride in view of the work of Fast[14] already mentioned.

Since manganese can diffuse at the testing temperatures used, it seems logical to explain the observed secondary strain-age-hardening phenomena by the precipitation of manganese carbide in the dislocations. Above about 200° C, the iron carbide dislocation precipitates become increasingly unstable as the temperature increases and tend to go into solution, so that the stress for a given strain decreases rapidly (Fig. 3). When manganese is present, however, the dislocations are favourable sites for the precipitation of stable manganese carbide, if the temperature is high enough, so that the rate of work-hardening is increased. Since there is ample carbon available in solution, the controlling factor will be the rate of diffusion of manganese, the amount available, and the activation energy of formation of the alloy carbide. The evidence suggests that the maximum effect is obtained at about 300–400° C. At still higher temperatures even the manganese carbide becomes unstable and the stress falls in a way similar to that of a low-manganese steel at a lower temperature. Other similarities may be noted, for example, when manganese is added a second minimum in reduction of area is observed at about 500° C. Also, the fluctuations of stress observed during straining at about 100–200° C are again observed at 200–300° C when manganese is present.

Although many of the characteristics of the 200° C maximum are reproduced in the second maximum, two points of difference were noted. In the first instance aluminium deoxidation has no effect on the maximum in stress due to manganese, whereas it reduces the stress of the maximum at 200° C. This might have been expected since at 200° C the maximum in stress is related to the total amount of available interstitial solute, whereas at 300–400° C the maximum is related to the amount of available manganese and not to the amount of interstitial solute, of which there is an ample supply at this temperature. Thus the presence or absence of nitrogen is not important. This contrasts with creep at high temperature where nitrogen is important.

Secondly, although a sharp change in the rate of work-hardening at the maximum load was noted in the tests at about 200° C, a more gradual change was observed at 300–400° C. The reason for this is similar to that given above. At 200° C, precipitation proceeds rapidly in the dislocations as soon as they form, but stops rather abruptly as the supply of interstitial solute gives out, so that the rate of work-hardening up to the maximum load is high and the elongation at the maximum load is small. At 300–400° C the controlling factor is not the number of dislocations and the supply of carbon and nitrogen, but the rate of diffusion of manganese and the activation energy of formation of the carbide. Thus, at this temperature, the rate of change of work-hardening is more gradual and the elongation at the maximum load tends to increase rather than decrease.

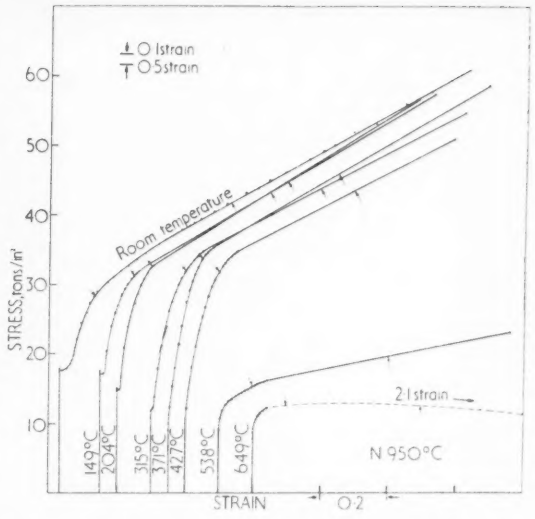

Fig. 12—True stress/strain curves for steel 7 (0·12% C, 1·35% Cr)

On increasing the manganese to 3·7% a considerable increase in strength results. It is interesting to note that the general effect is to raise the strength at a very small strain, but that the rate of work-hardening thereafter is not much different from that of the softer steels. This would appear to indicate that dispersion-hardening raises the yield point or proof stress but does not greatly influence the rate of work-hardening. This aspect of the subject is, however, outside the scope of this paper.

CHROMIUM

The effect of chromium was studied by testing five casts of steel, two being deoxidized with aluminium. Details of these casts are given below:

Steel	Type	C, %	Si, %	Mn, %	Cr, %	Al Addition, lb/ton
1	H.F.	0·11	0·07	0·07	0·01	0·5
6	H.F.	0·12	0·06	0·13	0·66	0·5
7	H.F.	0·12	0·02	0·11	1·35	0·5
8	H.F. *	0·11	0·15	0·36	2·61	0
1A	H.F.	0·105	0·10	0·15	0·01	3
6A	H.F.	0·11	0·09	0·18	0·63	3
7A	H.F.	0·10	0·15	0·12	1·41	3

*6 cwt. H.F. furnace

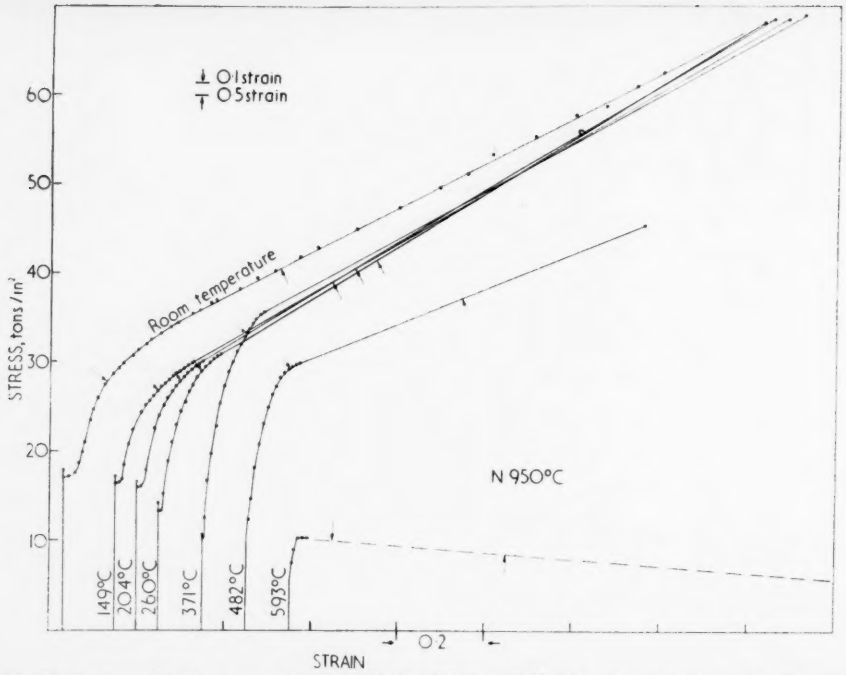

Fig. 13—True stress/strain curves for steel 7A (0·10% C, 1·40% Cr) deoxidized with aluminium

The results obtained from steels *6* and *6A* were not markedly different from those of the comparable manganese steel. The true stress/strain curves for steels *7* and *7A* are shown in Figs. 12 and 13, and those for steel *8* in Fig. 14. Iso-strain curves for steels *6* and *6A* are shown in Fig. 15, and those for steels *7* and *7A* in Fig. 16. The effect of chromium is summarized by the iso-strain curves in Fig. 17. The variation of reduction in area with temperature is shown in Fig. 18.

The addition of about 1·35% chromium has less effect on the strength at room temperature than the addition of 1·5% manganese. Indeed it is not very much stronger than steel *1* with no manganese or chromium. As shown in Fig. 12, the degree of work-hardening is not very pronounced at 200° C, but at a somewhat higher temperature the rate of work-hardening increases, as was the case with the 1·5%

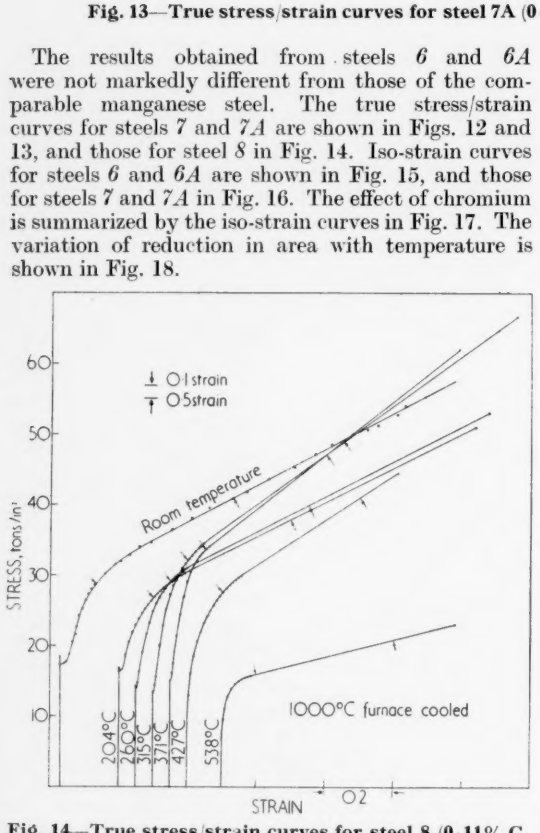

Fig. 14—True stress/strain curves for steel 8 (0·11% C, 2·6% Cr)

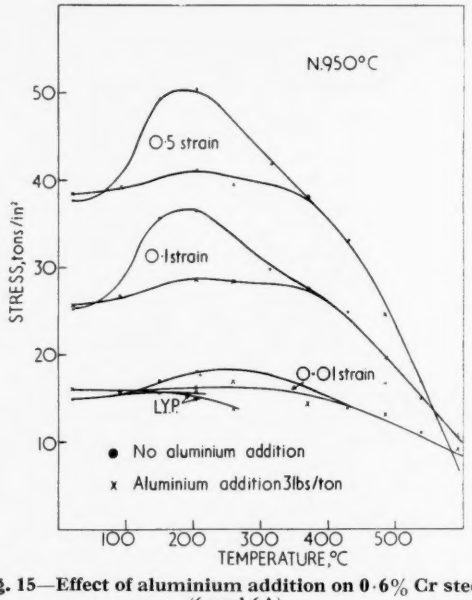

Fig. 15—Effect of aluminium addition on 0·6% Cr steels (6 and 6A)

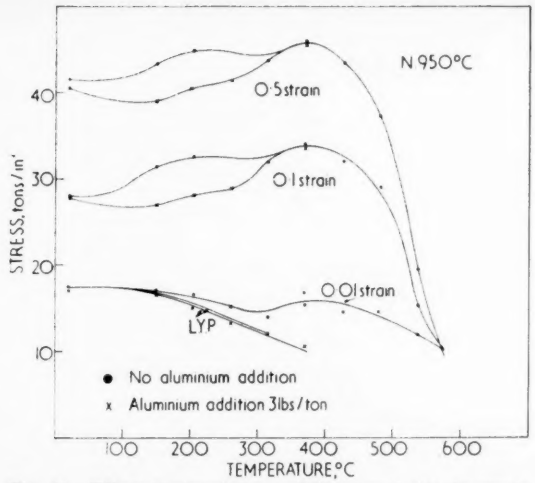

Fig. 16—Effect of aluminium addition on 1·4% Cr steels (7 and 7A)

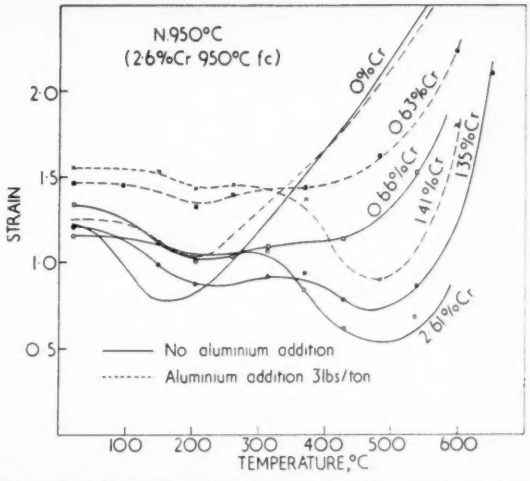

Fig. 18—Effect of chromium on ductility (steels 1, 1A, 6, 6A, 7, 7A, and 8)

manganese steel. The addition of aluminium (Fig. 13) decreases the strength at about 200° C, but has no effect on the strength at higher temperatures. The 2·6% chromium steel (Fig. 14) was tested in the furnace-cooled condition, and at room temperature the true stress/strain curve is much the same as that of the 1·35% chromium steel in the normalized condition. The furnace-cooled condition was chosen simply because it is more usual to use this treatment with such steels, and it also serves as a comparison later with a 2¼% chromium steel containing molybdenum. A notable feature of the results of steel *8* (Fig. 14) is that there appears to be no strain-age-hardening at about 200° C though strain-age-hardening is pronounced at a higher temperature.

The 0·6% chromium steels are compared in Fig. 15. No comment is required except that the results are very similar to those of the corresponding manganese steel. The 1·3% chromium steels are compared in Fig. 16; as will be noted there is only a very small increase in stress at 200° C even with steel 7. A pronounced maximum at about 300–400° C is obtained. Within experimental error this is about the same temperature as, if not a little higher than, the maximum obtained from manganese. It should be noted from Fig. 16 that the yield point persists to almost 400° C, a much higher temperature than was the case of the other steels discussed.

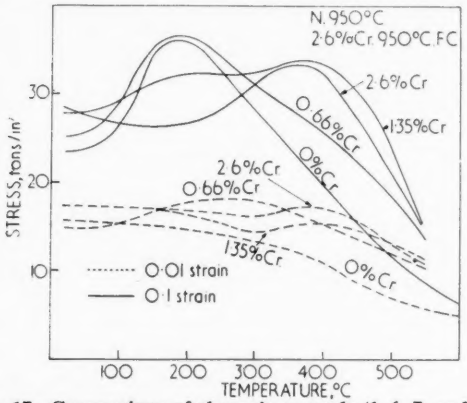

Fig. 17—Comparison of chromium steels (1, 6, 7, and 8) at 0·01 and 0·1 strain

Figure 17 clearly shows the strengthening effect of chromium at 300–400° C, but perhaps a more interesting point is its effect on the maximum at 200° C. With increasing chromium the maximum rapidly decreases, and with 2·6% chromium has apparently disappeared. As shown in Fig. 18, the reduction of area curves for the chromium steels are similar to those of the manganese steels except that at 200° C the minimum in reduction of area is less pronounced.

As indicated by Austin[15] the carbides formed by chromium in steels containing less than 3% chromium are very similar to those of manganese. It is, therefore, perhaps not surprising that the precipitation of chromium carbide in the dislocations has an effect very similar to that of the precipitation of manganese carbides and occurs at much the same temperatures. The previous explanation given for manganese thus serves also to explain the effect of chromium at 300–400° C. The effect of chromium on the maximum in stress at about 200° C is also similar to the effect of manganese, except that it is much more pronounced. It appears that the carbon and nitrogen atoms in interstitial solution at the lower temperatures are very firmly fixed at "chromium sites," i.e. interstices adjacent to chromium atoms. As a result, with sufficient chromium present little or no carbon or nitrogen is available for strain-age-hardening. Some effect still remains, as evidenced by the slight minimum in reduction of area which still persists even with the 2·6% chromium steels. This also explains the very high temperature at which a yield point is still observed with the chromium steels. As already indicated, the extension at the yield point decreases with increasing temperature because precipitation in the dislocations augments the normal strengthening effect of work-hardening. With chromium present, no precipitation occurs at about 200° C,

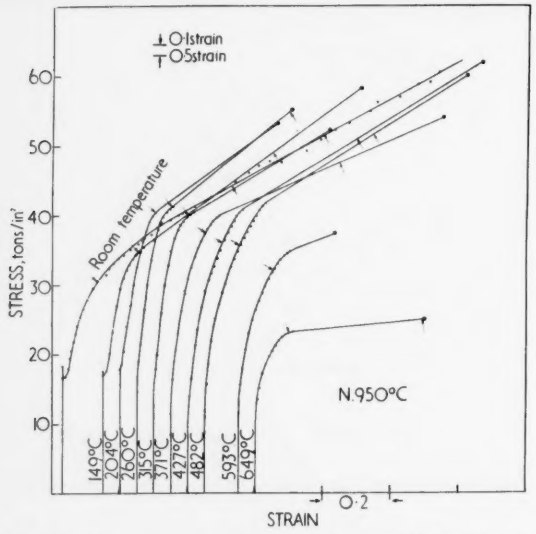

Fig. 19—True stress/strain curves for steel 10 (0·095% C, 1·46% Mo)

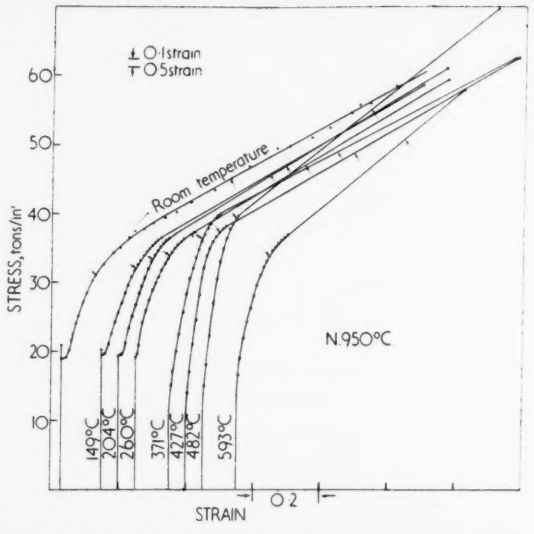

Fig. 20—True stress/strain curves for steel 10A (0·095% C, 1·53% Mo) deoxidized with aluminium

and thus the yield point is still observed at high temperatures until eventually it is eliminated by precipitation of chromium carbide during straining and by normal thermal effects.

MOLYBDENUM

Four molybdenum steels were tested, two being deoxidized with aluminium. Details are shown below:

Steel	Type	C, %	Si, %	Mn, %	Mo, %	Al Addition, lb/ton
1	H.F.	0·11	0·07	0·07	0·01	0·5
9	H.F.	0·105	0·16	0·11	0·505	0·5
10	H.F.	0·095	0·05	0·11	1·46	0·5
1A	H.F.	0·105	0·10	0·15	0·01	3
9A	H.F.	0·11	0·08	0·16	0·51	3
10A	H.F.	0·095	0·09	0·18	1·53	3

The true stress/strain curves for steels *9* and *9A* are similar to those of 0·5% manganese, except that the second strain-age-hardening effect occurs at a higher temperature. The curves for steels *10* and *10A* are shown in Figs. 19 and 20. Iso-strain curves for the 0·5% molybdenum steels are shown in Fig. 21, and for the 1·5% molybdenum steels in Fig. 22. The curves for reduction of area are given in Fig. 23.

As shown in Fig. 19, pronounced strain-age-hardening occurs with the 1·5% molybdenum steel at about 200° C. At a slightly higher temperature the strength decreases slightly, but at a still higher temperature marked hardening occurs. With the aluminium-killed steel (Fig. 20), very little hardening occurs at 200° C but at higher temperatures it is pronounced, the true stress/strain curves being little different from those of the steel without aluminium.

The iso-strain curves (Figs. 21 and 22) indicate that molybdenum causes a secondary strain-age-hardening effect with a maximum at about 500° C. As before, aluminium has no effect at the higher temperatures. It is interesting to note that, unlike chromium, quite a pronounced maximum in stress is obtained at 200° C in the steels containing molybdenum.

The effect of molybdenum is clearly demonstrated by comparing Figs. 21 and 22 with Fig. 3, so that a summary graph is not necessary. The molybdenum steels show a minimum in reduction of area at about 600° C (Fig. 23). It should be noted that the minimum at about 200° C is still quite pronounced, even with the 1·5% molybdenum steel. With increasing molybdenum this minimum occurs at a higher temperature of testing.

It is well known that molybdenum steels do not soften as readily on tempering as manganese steels, indicating either that the rate of diffusion of molybdenum is less than that of manganese or that molybdenum carbide is more difficult to form. It is to be expected, therefore, that the maximum in stress due

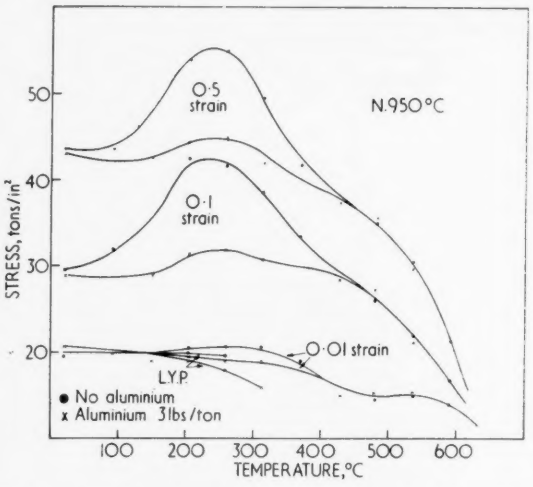

Fig. 21—Effect of aluminium addition on 0·5% Mo steels (9 and 9A)

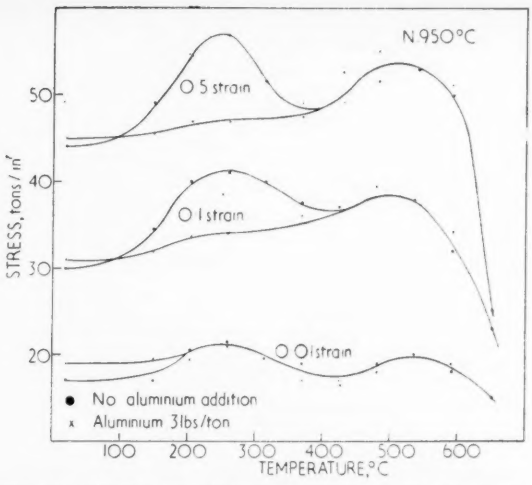

Fig. 22—Effect of aluminium addition on 1·5% Mo steels (10 and 10A)

to precipitation of molybdenum carbide in the dislocations will occur at a higher temperature than the maximum due to manganese. Apart from this the same argument as used for manganese explains the effect of molybdenum.

The effect of molybdenum on the maximum in stress at 200° C and the minimum in reduction of area at the same temperature is more difficult to explain. Molybdenum has a greater affinity for carbon than chromium and thus it might have been expected that molybdenum would tend to eliminate the peak at 200° C, as was the case with chromium, whereas in fact molybdenum has very little effect. On the other hand it should be remembered that the molybdenum atom is almost twice as heavy as the chromium atom, so that a 1·5% molybdenum steel contains approximately the same number of molybdenum atoms as there are chromium atoms in a 0·75% chromium steel. Since the 0·65% chromium steel shows a pronounced maximum at 200° C, the maximum in the 1·5% molybdenum steel is thus partly explained. As already mentioned, the interstitial solute atoms tend to migrate into interstices adjacent to manganese, chromium, and probably also molybdenum

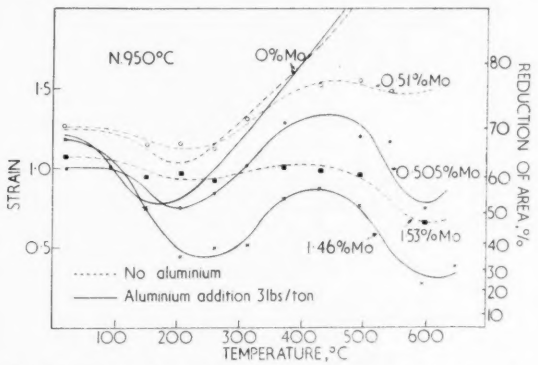

Fig. 23—Effect of molybdenum on ductility (steels 1, 1A, 9, 9A, 10, and 10A)

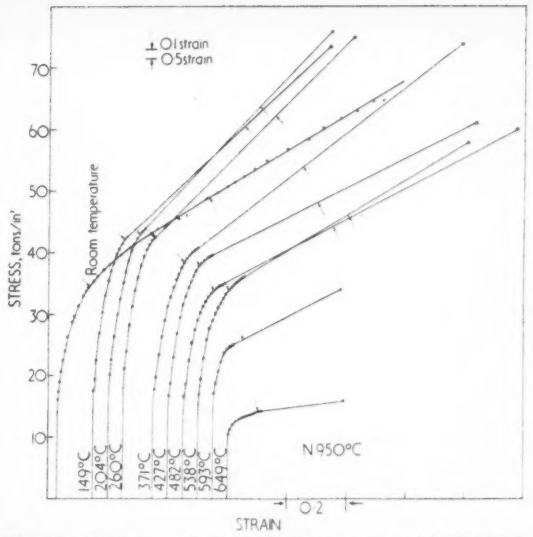

Fig. 24—True stress/strain curves for steel 13 (0·085% C, 3·14% W)

atoms. But the number of interstices adjacent to such atoms is proportional to the number of alloy atoms present and not to the actual weight per cent. Thus the amount of interstitial solute which can be fixed at manganese, chromium, or molybdenum sites is also proportional to the number of such atoms present. On this argument, therefore, the effect of molybdenum is brought more into line with the effect of chromium and manganese. It must be admitted, however, that the above is only a partial explanation, since even allowing for atomic weight the effect of molybdenum is less than might be expected. If molybdenum increases the solubility of carbon and nitrogen in steel, then the effect would be fully explained, since more interstitial solute would thus be available for precipitation in dislocations at 200° C. The author has reason to suppose that this is correct, but prefers to put it forward only as a possibility at this stage.

It is interesting to note that Edwards[4] also found in tests at room temperature that the same weight of molybdenum is not as effective as manganese or chromium in suppressing strain-age-hardening.

TUNGSTEN

Three tungsten steels were tested. Since these contained about 0·4% manganese, they can be compared with steel *2*. Details of these steels are given below:

Steel	Type	C, %	Si, %	Mn, %	W, %	Al Addition, lb/ton
2	**B.O.H.**	**0·10**	**0·05**	**0·43**	**0**	**0·5**
11	**H.F.**	**0·14**	**0·09**	**0·48**	**0·50**	**0**
12	**H.F.**	**0·125**	**0·09**	**0·46**	**1·11**	**0**
13	**H.F.**	**0·085**	**0·10**	**0·34**	**3·14**	**0**

The results obtained from steels *11* and *12* are not markedly different from those of steel *2*, except that the former are slightly stronger at about 500° C. The true stress/strain curves for steel *13* are shown in Fig. 24. Iso-strain curves for all three tungsten steels are shown in Fig. 25. The curves of reduction in area are shown in Fig. 26.

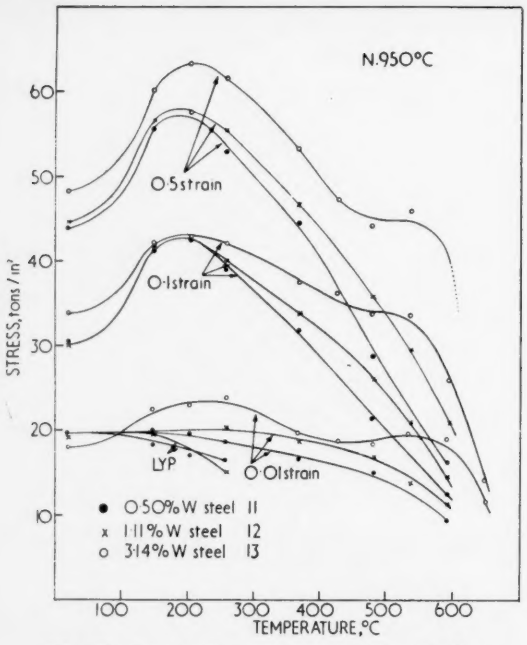

Fig. 25—Iso-strain curves on tungsten steels

From Fig. 24 it is evident that strain-age-hardening occurs at about 200° C and that hardening is quite pronounced, even at temperatures above 500° C. The iso-strain curves in Fig. 25 show this more clearly. Tungsten causes strain-age-hardening with a maximum effect at about 500–550° C. The maximum in stress at 200° C is pronounced in all three steels.

Tungsten forms carbides very similar to that of molybdenum but since the tungsten atom is twice as heavy as molybdenum, twice as much tungsten is required to give the same amount of carbide. By comparing the tungsten and molybdenum steel, it is evident that the 1% tungsten steel behaves somewhat similarly to the 0·5% molybdenum steel. The 3% tungsten steel is, however, not quite as strong at about 500° C as the 1·5% molybdenum steel indicating that some other effect influences the results. This is possibly the rate of diffusion since tungsten is

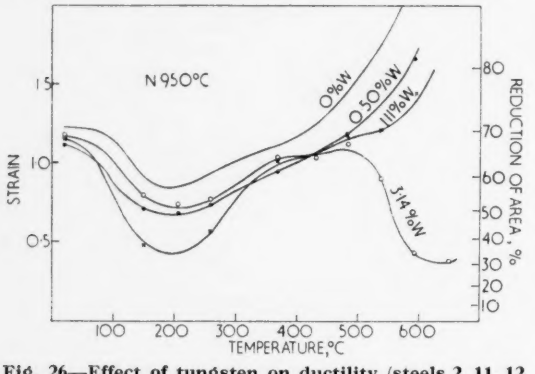

Fig. 26—Effect of tungsten on ductility (steels 2, 11, 12 and 13)

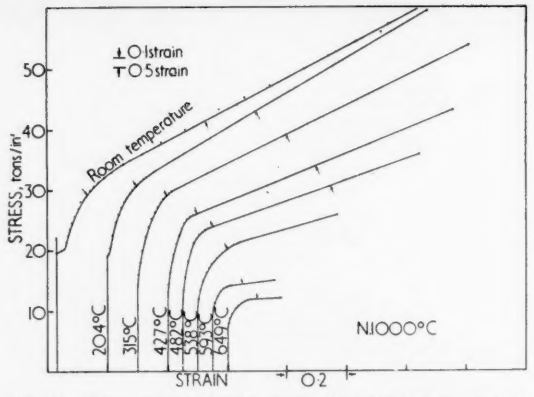

Fig. 27—True stress/strain curves for steel 14 (0·10% C, 0·25% V)

more sluggish than molybdenum. Like molybdenum, tungsten causes a minimum in reduction of area at about 600° C. The arguments used to explain the effect of molybdenum thus apply equally well to tungsten.

It is interesting to note that the effect of the 0·4% manganese present in these steels is reflected in the results. In Fig. 25 it tends to raise the stress between 300–500° C. Similarly in Fig. 26 it flattens the reduction in area curves at about the same temperature.

VANADIUM

Five vanadium steels were tested, two being normalized at a slightly higher temperature. These steels contained about 0·4% manganese and are compared with steel 2. The details of the steels are given below:

Steel	Type	Treatment	C, %	Si, %	Mn, %	V, %	Al Addition, lb/ton
2	H.F.	N. 950° C	0·10	0·05	0·43	0	0·5
14	H.F.	N.1000° C	0·10	0·10	0·65	0·25	0
15	H.F.	N.1000° C	0·12	0·08	0·63	0·55	0
16	H.F.	N.1000° C	0·13	0·09	0·48	1·14	0
17	H.F.	N.1050° C	0·10	0·11	0·47	0·25	0
18	H.F.	N.1050° C	0·10	0·10	0·42	0·55	0

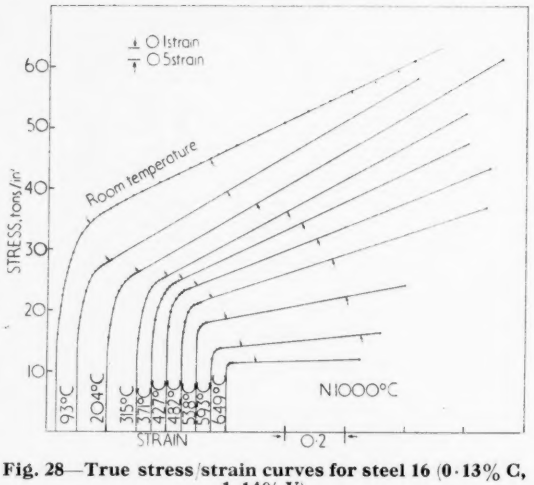

Fig. 28—True stress/strain curves for steel 16 (0·13% C, 1·14% V)

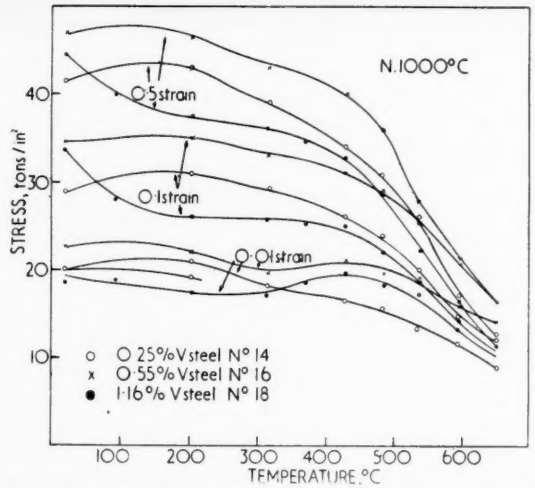

Fig. 29—Effect of vanadium on high-temperature strength

The true stress/strain curves for steel *14* are shown in Fig. 27, and for steel *16* in Fig. 28. Iso-strain curves for steels, *14*, *15*, and *16* are shown in Fig. 29, and for steels *17* and *18* in Fig. 30. The curves of reduction in area are given in Fig. 31.

Figure 27 shows that slight hardening occurs at about 200° C. Thereafter the strength decreases more or less gradually with increasing temperature. Figure 28 shows softening at 200° C and a further gradual decrease in strength with increasing temperature.

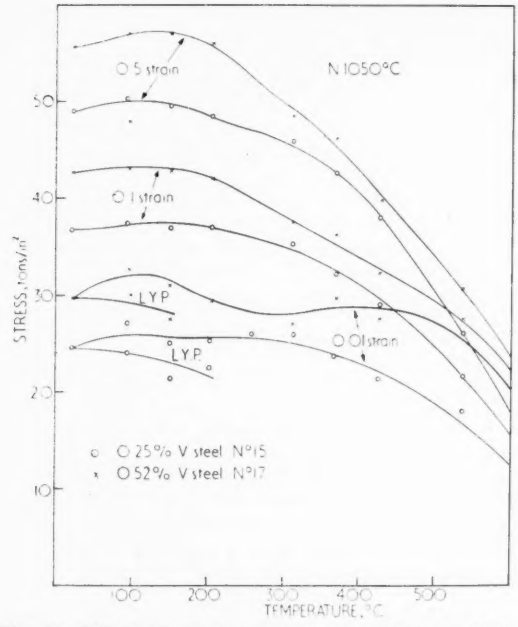

Fig. 30—Effectof 0·25 % and 0·5% vanadium on high-temperature strength

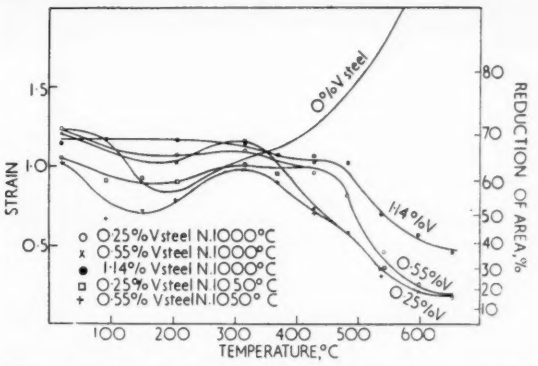

Fig. 31—Effect of vanadium on ductility of 0·4% manganese steel

The iso-strain curves in Fig. 29 show that there is a slight maximum in stress at about 200° C for the 0·25 and 0·5% vanadium steels, but that the 1% vanadium steel softens markedly between room temperature and 200° C. A further maximum which is particularly evident at 0·01 strain occurs at about 400° C. As shown in Fig. 30, by normalizing at 1050° C the maximum in stress at about 200° C is slightly increased but the strength at higher temperatures is not affected appreciably.

The reduction of area curves (Fig. 31) are interesting. At 200° C a minimum occurs with the 0·25% and 0·5% vanadium steels but there is no minimum with the 1% vanadium steels. Between 300° and 500° C there is a slight minimum or flattening of the curves, evidently related to the maximum in stress at about the same temperature. Finally there is a pronounced minimum in reduction of area at about 650° C which is not related to any definite maximum in stress.

Vanadium has a very strong affinity for carbon and nitrogen, and vanadium heat-treated steels require a very high temperature for softening. Indeed such steels very often show a precipitation-hardening effect as iron carbide dissolves and is replaced by vanadium carbide in fine dispersion. The rate of diffusion of vanadium is thus slower even than that of molybdenum or tungsten. However, the above steels have little alloy content apart from vanadium, and since vanadium raises the transformation temperature markedly, the transformation on cooling occurs at a high temperature so that most of the carbon is precipitated as massive vanadium carbide and the steels are soft.

Vogel and Martin[16] claim that vanadium increases the solubility of carbon in ferrite and the author considers this quite probable. As already indicated, it is probable that molybdenum also increases the solubility of carbon in ferrite. As regards nitrogen, however, vanadium nitride appears to be so stable that very little nitrogen remains in solution below about 1000° C. In this respect vanadium behaves like aluminium. It appears to be the presence of aluminium nitride or vanadium nitride particles which prevents grain-coarsening of austenite. Above a certain temperature when the nitride goes into solution a sudden increase in grain size is obtained.

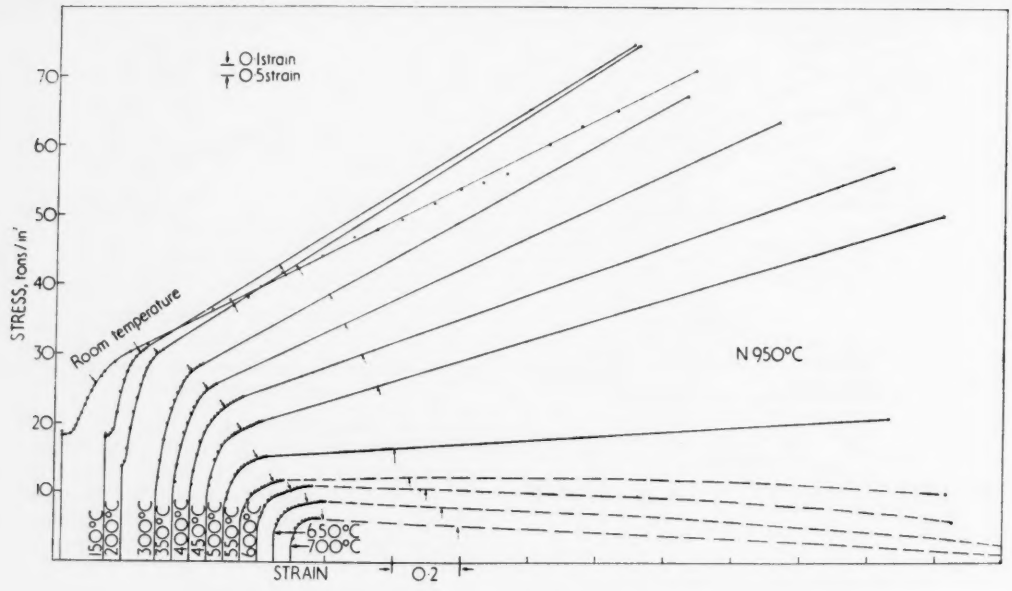

Fig. 32—True stress/strain curves for steel 20 (0·075% C, 0·28% Ti)

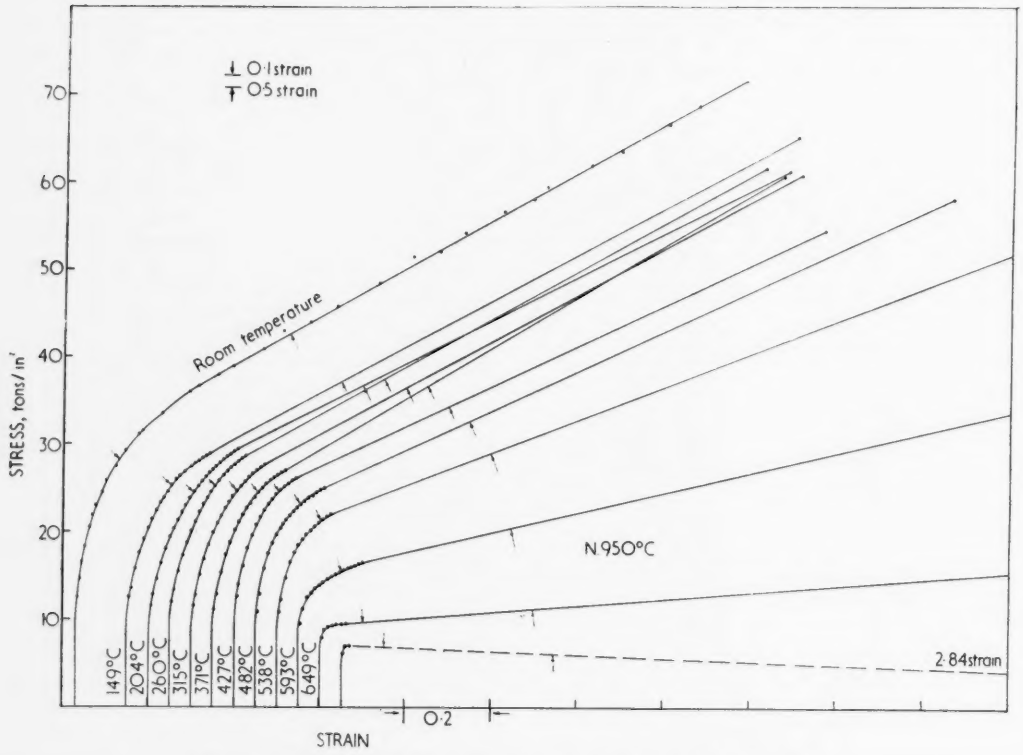

Fig. 33—True stress/strain curves for steel 22 (0·06% C, 1·44% Ti)

The carbon remaining in solution in the vanadium steel is, however, very strongly fixed in interstices adjacent to vanadium atoms, so that only a small maximum in stress is obtained for the 0·25 and 0·5% vanadium steels. With the 1·14% vanadium steel all the carbon is fixed and the stress decreases between room temperature and 200° C, as is to be expected with an increase in temperature of testing in the absence of strain-age-hardening. With increasing temperature, however, the carbon atoms are less rigidly fixed so that some become available for precipitation in the dislocations. The temperature is too high for iron carbide to be effective, but manganese carbide is stable at about 400° C, and so the second maximum in stress at about 400–500° C is probably due to precipitation of manganese carbide in the dislocations. The slight minimum in reduction of area at about 400° C supports this contention. The temperature of the maximum is somewhat higher than that obtained with the manganese steels and appears to be related to the presence of vanadium, which inhibits the resolution and spheroidization of manganese carbide. Further evidence to prove this has been obtained by testing combinations of alloying elements. It suffices to state that at present the above effect appears to be a general one and that one alloying element modifies the effect of another.

Figure 31 shows a third minimum in reduction of area at about 650° C which must be attributed to the presence of vanadium. However, no noticeable maximum in stress occurred at about the same temperature. It may be that the rate of straining was too rapid to allow time for appreciable diffusion of vanadium and so no maximum is observed. On the other hand, a still slower rate of straining would allow time for creep to occur and would complicate the interpretation. It should be remembered also that most of the available carbon is present as massive vanadium carbide which has little tendency to go into solution to feed the dislocation precipitates. The strengthening effect of vanadium is more clearly shown on steels containing other alloying elements, and when there are appreciable amounts of other carbides which can provide carbon for the precipitation of vanadium carbide in the dislocations. It is proposed to discuss this aspect more fully in a later paper.

As shown in Fig. 30, the effect of increasing the normalizing temperature is to increase the strength at the lower temperatures. This could be attributed to the solution of some nitride on normalizing at the higher temperature.

TITANIUM

To investigate the effect of titanium, four steels were tested, each containing approximately 0·4% manganese. To minimize loss of titanium during manufacture these steels were deoxidized with aluminium. They can thus be compared with steel *2A*. The details of the steels are given below:

Steel	Type	Treatment	C, %	Si, %	Mn, %	Ti, %	Al Addition, lb/ton
2A	B.O.H.	N. 950° C	0·10	0·06	0·45	0	3
19	B.O.H.	N.1150° C W.Q.1150° C	0·075	0·07	0·35	0·22	2
20	B.O.H.	N. 950° C	0·075	0·07	0·35	0·28	2
21	E.F.	B. 950° C	0·07	0·35	0·46	0·55	2
22	H.F.	N. 950° C	0·06	0·45	0·35	1·44	2

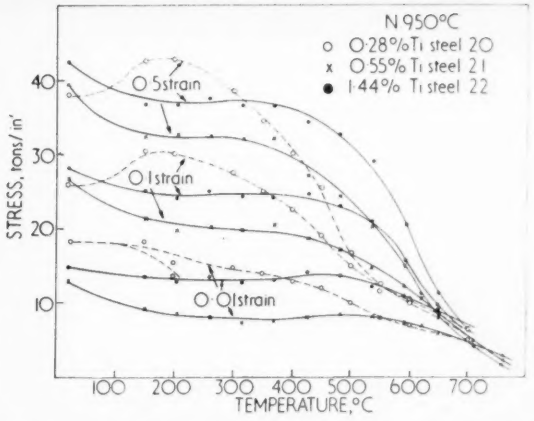

Fig. 34—Effect of titanium on the high-temperature strength of 0·4% manganese steel

The true stress/strain curves for steel *20* are shown in Fig. 32. The curves for steels *21* and *22* are similar and only those of the latter steel are reproduced in Fig. 33. Iso-strain curves of steels *20*, *21*, and *22* are shown in Fig. 34. The true stress/strain curves for steel *19* are shown in Figs. 35 and 37. The effect of variation in heat treatment is shown by the iso-strain curves in Fig. 36. All the curves of reduction in area are shown in Fig. 38.

Figure 32 shows that slight hardening occurs at *c.* 200° C with the 0·28% titanium steel; thereafter the stress falls gradually with increasing temperature of testing. The curves for the 1·44% titanium steel (Fig. 33) are similar to those of the high-vanadium steel. There is a rapid decrease in strength between room temperature and about 200° C, and a slow decrease thereafter. The iso-strain curves for steel *20* in Fig. 34 show a maximum in stress at about 200° C and also a bulge in the stress curves at higher temperature similar to that obtained with 0·5% manganese steel. The higher titanium steels show no maximum at 200° C, the stress falling rapidly between room temperature and 200° C. At higher temperatures a slight maximum is obtained at about 500° C. This is quite clear at 0·01 strain.

Figure 35 shows that for the 0·22% titanium steel normalizing at 1150° C, the strength decreases at about 100° C and then increases to a slight maximum at about 200° C. At higher temperatures the strength falls off slowly. Water-quenching from 1150° C (Fig. 37) greatly increases the strength at small strains and hardening occurs at about 200° C. At higher temperatures the strength decreases. The iso-strain curves in Fig. 36 show the effect of heat treatment more clearly. The curves for steel *20* normalized at 950° C are included. Although at temperatures up to about 400° C steel *20* normalized at 950° C is stronger than steel *19* normalized at 1150° C, at higher temperatures it is much weaker. The high-temperature treatment causes a maximum in stress at about 500–600° C in steel *19* which is not shown by steel *20* normalized at 950° C.

A minimum in reduction of area was obtained

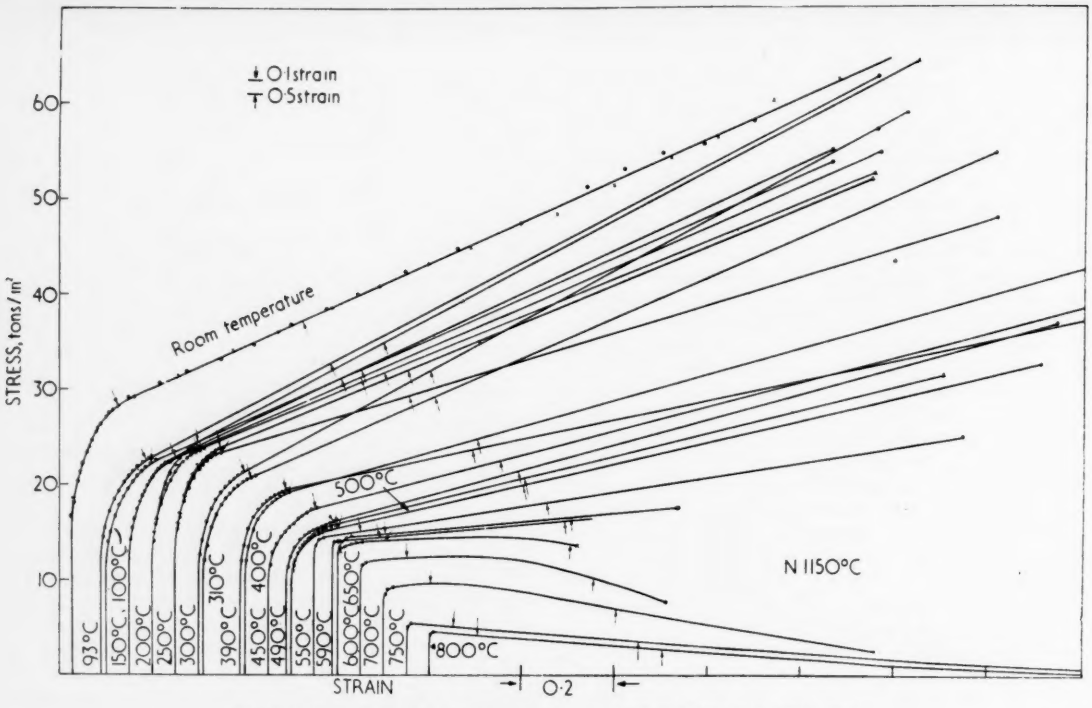

Fig. 35—True stress/strain curves for steel 19 (0·075% C, 0·22% Ti)

for all the steels at about 200–350° C. With the lower-titanium steel this minimum occurs at about 200° C, but shifts to a higher temperature as the titanium increases. A second minimum in reduction of area occurs at about 600° C. For the steels normalized at 950° C, the minimum is more pronounced for the 0·55% titanium steel than is the case with steel of lower and higher titanium content. The low-titanium steel heat-treated at 1150° C gave a very pronounced minimum at about 600° C. Unlike the vanadium steels, there appears to be no third minimum in reduction of area. It should be pointed out, however, that the reduction of area at 800° C for the low-titanium steel heat-treated at 1150° C is less than might be expected from the trend of the other curves. It seems possible, therefore, that the third minimum in reduction of area occurs at 750–800° C but is masked by other phenomena.

Titanium has a greater affinity for carbon than vanadium and raises the transformation temperature rapidly. In fact, the normalizing treatment at 950° C is simply a tempering treatment for the 0·55% and 1·44% titanium steels, since the change points, if any, occur at a higher temperature. For this reason any carbon out of solution in the steels normalized at 950° C will be in the form of massive titanium carbide. Like the steels containing vanadium or molybdenum, however, it is probable that some carbon remains in solution at room temperature, but this will be strongly fixed at titanium ' sites.' In fact, according to the published iron–carbon–titanium diagram,[17] about 0·1% carbon is in solution at room temperature in a 1% titanium steel. Thus, for the 1·44% titanium steel, it is possible that most if not all of the carbon present is in interstitial solution.

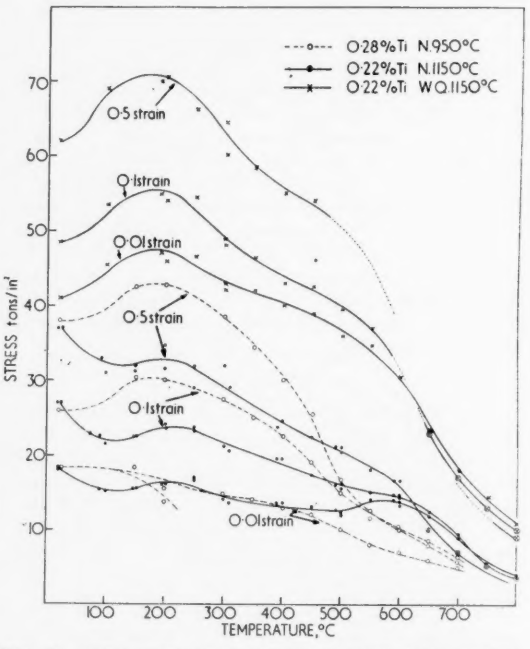

Fig. 36—Effect of heat treatment on high-temperature strength of low-titanium steels (19 and 20)

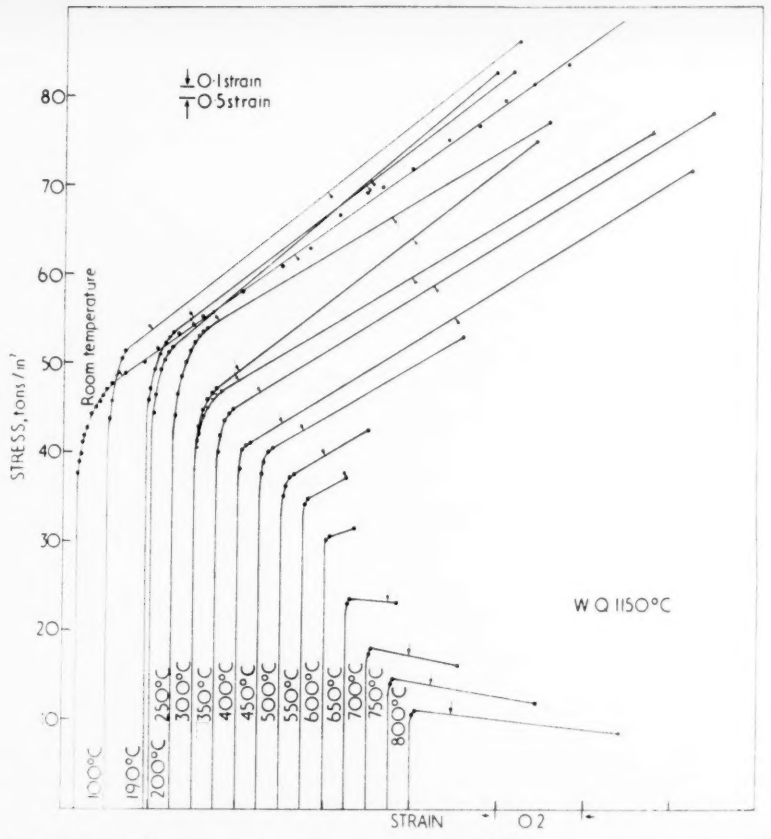

Fig. 37—True stress/strain curves for steel 19 (0·075% C, 0·22% Ti)

Titanium nitride is very insoluble and it is probable that no nitrogen remains in solution at the testing temperature used. In any case these steels were deoxidized with aluminium and this of itself would remove most of the nitrogen from interstitial solution. Since the vanadium steels which were not deoxidized with aluminium behave very like the titanium steels, this is indirect evidence that vanadium acts like aluminium in removing nitrogen from solution.

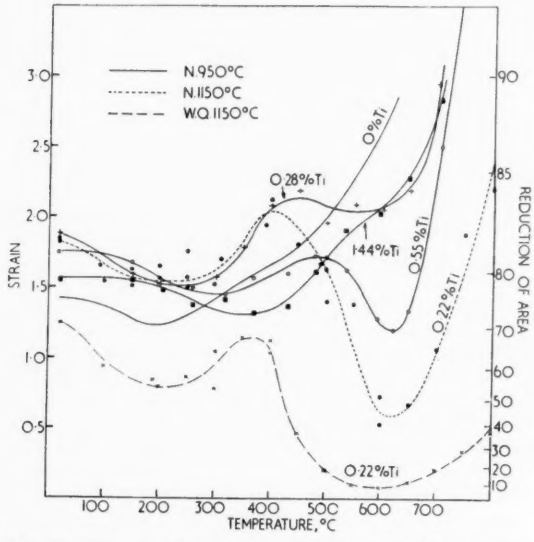

Fig. 38—Effect of titanium on ductility of 0·4% manganese steel

Some of the titanium in the 0·28% titanium steel *20* is probably present as carbide, but in any case there is not sufficient titanium remaining to fix all the carbon in solution at interstices adjacent to titanium atoms. As a result steel *20* behaves somewhat like a simple manganese steel, that is, it gives a maximum in stress at about 200° C and an increase in stress at higher temperature due to the manganese present. However, recovery or recrystallization of the test piece on straining begins to occur at temperatures above about 500° C, as shown by a lower stress at fracture than at the maximum load. In the absence of this effect, the influence of manganese might persist to a higher temperature, as indeed is the case when the recrystallization temperature is raised by a different heat treatment. This is discussed more fully later. Some evidence of the effect of manganese persisting to higher temperatures is given by the fact that the second minimum in reduction of area for steel *20* occurs at a higher temperature than for the vanadium steels.

With 0·55% or 1·44% titanium the interstitial carbon, even though plentiful, is effectively fixed at titanium 'sites,' so that no maximum in stress occurs at 200° C. Indeed the stress decreases between

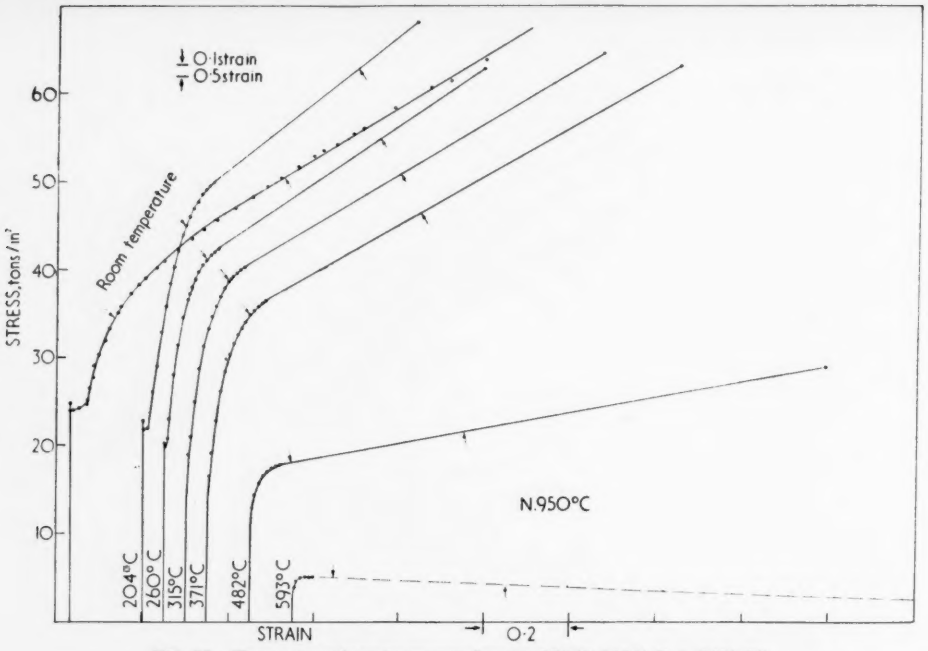

Fig. 39—True stress/strain curves for steel 23 (0·115% C, 1·04% Si)

room temperature and 200° C, as in the 1·14% vanadium steel. At higher temperatures thermal agitation tends to free some of the carbon fixed at titanium ' sites ' and the behaviour is similar to that of the 1·14% vanadium steel. A maximum in stress occurs at about 500° C, which, as before, can be attributed to the precipitation of manganese carbide in the dislocations formed by straining. Associated with this maximum in stress is a minimum in the reduction of area. With the 1·44% titanium steel this minimum has degenerated to an inflexion in the curve and the reduction of area is high. With 0·55% titanium the reduction of area falls to a low value, and with 0·22% titanium to a still lower value. This latter steel was, however, heat-treated at 1150° C. If it had been normalized at 950° C like the 0·28% titanium steel, recrystallization would have occurred above about 500° C and increased the reduction of area. Thus, it appears that with increasing titanium content the intensity of the minimum in reduction of area due to manganese is decreased. The vanadium steels show a similar effect (Fig. 31). This is a further example of the effect of one alloying element modifying the effect of another. More evidence to prove this is available but discussion of it is deferred to a subsequent paper.

Since about 1% titanium suffices to suppress the change points in low-carbon steel, the 1·44% titanium steel is non-heat-treatable. With 0·55% titanium complex structures are obtained on heat-treating at 1150° C and results are erratic. For this reason the effect of heat treatment was investigated using only the low-titanium steels (*19* and *20*). Both steels change to austenite at about 900° C but some titanium carbide is still out of solution. The temperature must be raised to at least 1100° C before a uniform austenite is obtained and all the carbide and probably nitride are in solution. On water-quenching from 1150° C, any carbide and nitride which forms will be that of iron, and a considerable amount of carbon will remain in supersaturated solution. Since under these conditions too much carbon is present in interstitial solution to be fixed by the small amount of titanium in steel *19*, a maximum in stress is obtained at about 200° C by precipitation of iron carbide during straining (Fig. 36). A second maximum in stress occurs at about 600° C, associated with a pronounced minimum in reduction of area. Since any effect due to titanium should occur at a higher temperature than that for vanadium (i.e. higher than 650° C) it seems logical to attribute the above phenomena at about 600° C to the influence of manganese.

Recovery of recrystallization sets in at about 600° C, and this has probably masked any effect due to titanium at higher temperature. As with vanadium, it appears as if titanium is only effective at high temperatures if sufficient other alloying elements are present. A few tests on the 0·28% titanium steel water-quenched from 1150° C were carried out and the results obtained were similar to those of the lower-titanium steel.

On normalizing at 1150° C, most of the titanium remains in solution and any carbide which forms on cooling will be iron carbide. However, unlike the water-quenched steel, there will be little or no supersaturation of carbon in solution. Thus only a small maximum in stress is obtained at 200° C (Fig. 36), since most of the carbon in solution is fixed at titanium ' sites.' This contrasts with the larger maximum in stress at 200° C obtained with the 0·28% titanium steel normalized at 950° C. However, as already explained, in this condition some of the titanium is

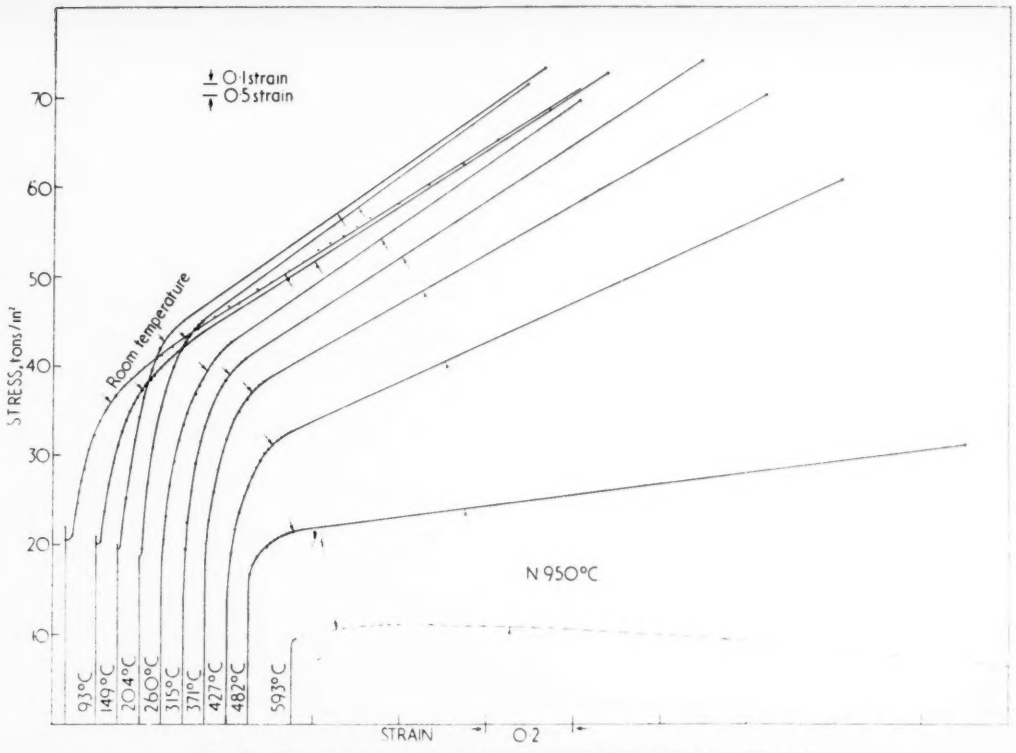

Fig. 40—True stress/strain curves for steel 24 (0·125% C, 3·08% Ni)

present as titanium carbide and less is available to fix the interstitial carbon.

Similarly, because of the higher titanium content of the ferrite on normalizing at 1150° C compared with 950° C, the recovery of recrystallization temperature is increased from about 500° C to about 600° C. Because of this, the effect of manganese is more definite in the steel normalized at 1150° C, giving a maximum in stress at 600° C and a pronounced minimum in reduction of area at about the same temperature. Some test pieces of the 0·28% titanium steel were also normalized at 1150° C. The results were erratic but similar to those obtained on water-quenching. For this reason they were not included.

It is clear from the above discussion, and also from the results of the steels containing other alloying elements, that the amount of carbon and nitrogen in interstitial solution at any temperature, and the proportion fixed at alloy sites, is one of the controlling factors in determining whether strain-age-hardening will occur and also its intensity. The development of methods for determining interstitial solubility is thus of importance if a quantitative measure of strength and strain-age-hardening is ever to be attempted. Internal friction methods are promising but so far have been used only on very simple material and at temperatures not far removed from room temperature.

SILICON AND NICKEL

So far all the alloying elements considered have been those which form stable carbides, and it has been postulated that these carbides can precipitate in the dislocations during straining. Thus the addition of alloying elements which do not form stable carbides or nitrides should not give rise to any new strain-age-hardening effects. To check this, a 1% silicon steel and a 3% nickel steel were tested. It was also thought worthwhile to consider a simple austenitic steel. For this purpose two steels, one containing 30% and the other 40% nickel, were prepared. Details of the steels are given below:

Steel	Type	C, %	Si, %	Mn, %	Ni, %	Al Addition, lb/ton
2	B.O.H.	0·10	0·005	0·43	0·07	0·5
23	H.F.	0·115	1·04	0·22	—	0
24	B.O.H.	0·125	0·12	0·56	3·08	1·5
25	H.F.	0·095	0·18	0·56	31·2	0
26	H.F.	0·09	0·18	0·48	41·6	0

The true stress/strain curves for steels *23* and *24* are shown in Figs. 39 and 40. Iso-strain curves for these steels are given in Fig. 41, and the curves of reduction in area in Fig. 42. The true stress/strain curves for steels *25* and *26* are almost identical and only those of steel *26* are given in Fig. 43. Iso-strain curves of these steels are given in Fig. 44 and the curves of reduction of area in Fig. 45.

Figures 39 and 40 show that the 1% silicon and 3% nickel steels are not noticeably different from a simple manganese steel. This is more clearly demonstrated in Fig. 41. Both steels show a maximum in stress at about 200° C and a maximum at higher temperatures due to manganese. It will be noted, however, that 3% nickel reduces the height of the maximum in stress at 200° C and it may be that nickel reduces the solubility of carbon and nitrogen in iron.

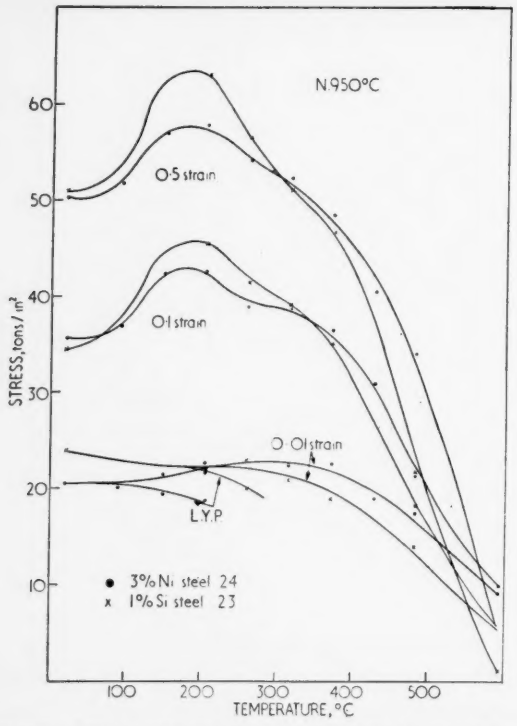

Fig. 41—Effect of 3% Ni and 1% Si on high-temperature strength

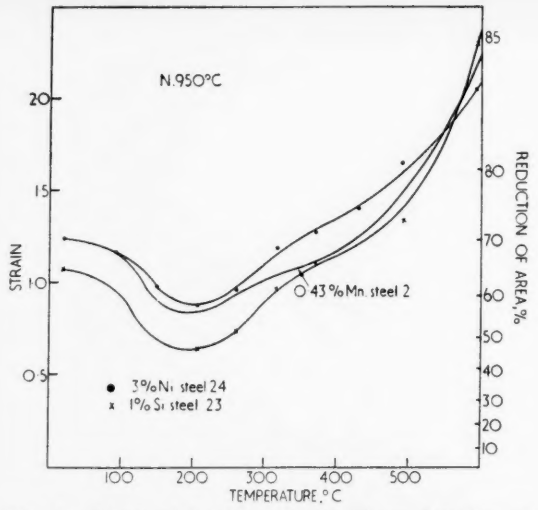

Fig. 42—Effect of 3% Ni and 1% Si on ductility of 0·4% Mn steel at high temperature

On the other hand, the high aluminium addition may have an influence even though the steel was coarse-grained in the McQuaid–Ehn test. Silicon, if anything, increases the maximum in stress and may increase the solubility of carbon and nitrogen in iron. There is, however, a possibility that silicon nitride can form, in which case the increase in stress may be due to precipitation of silicon nitride in the dislocations at about the same temperature as iron nitride. The curves of reduction in area (Fig. 42) show that nickel and silicon do not give rise to a new minimum in reduction of area, the curves being very similar to that of steel *2*.

The austenitic steel (Fig. 43) behaves very differently from the ferritic steels, the maximum load occurring only after about 50% extension. This is usually attributed to the different work-hardening characteristics of the face-centred cubic, but it must be remembered that all the carbon is in interstitial solution and the possibility remains that precipitation of carbide occurs during straining. The results shown in Fig. 43 lend support to this possibility. From this there appears no doubt that strain-age-hardening is taking place, with a maximum effect at about 200° C. But this is the same temperature as that at which iron carbide precipitates in the ferritic steels. For want of any other explanation it must be assumed that precipitation of iron carbide can occur in the dislocations formed in austenite during straining. Corresponding to this maximum in stress there is a minimum in reduction of area as shown in Fig. 45. However, a second minimum in reduction of area is also obtained which is not associated with a maximum in stress. Since the fracture was of the intercrystalline type, some precipitate must have formed at the grain boundaries during straining.

COPPER

As a final experiment for the present paper it was decided to investigate a steel in which a precipitate other than carbide or nitride occurred. The obvious choice was a copper-bearing steel. Two casts of 1·5% copper steel of low manganese content were prepared of the same nominal analysis. The test pieces from one cast were maintained at temperature for half-an-hour before commencing the test, whereas the other

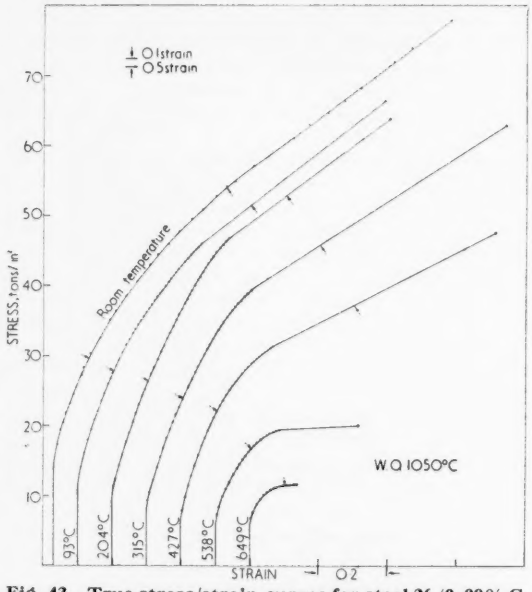

Fig. 43—True stress/strain curves for steel 26 (0·09% C, 41·6% Ni)

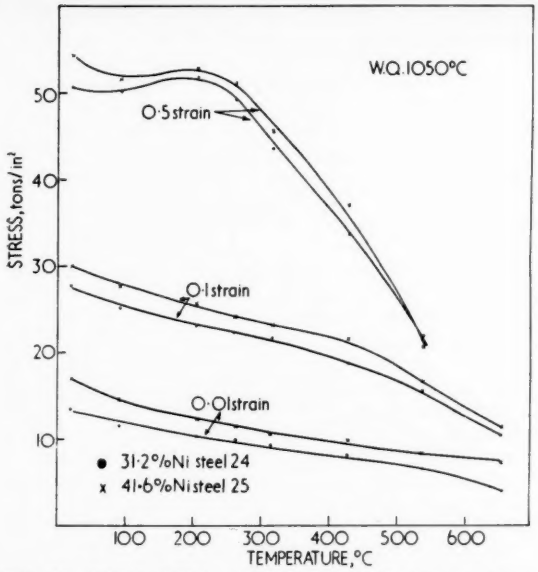

Fig. 44—Effect of 30% and 40% Ni on high-temperature strength

cast was held for one hour. Details of the steels are given below:

Steel	Type	C, %	Si, %	Mn, %	Cu, %	Al Addition, lb/ton
1	H.F.	0·11	0·07	0·07	0	0·5
27	H.F.	0·07	0·04	0·22	1·63	0·5
28	H.F.	0·08	0·05	0·13	1·48	0·5

The true stress/strain curves for both the copper steels are shown in Fig. 46 and iso-strain curves in Fig. 47. The curves of reduction in area are given in Fig. 48.

It will be noted from Fig. 46 that a high yield point is obtained at room temperature and that hardening occurs as usual at about 200° C. A second hardening effect appears to occur at a higher temperature. This is clearly shown in Fig. 47, a maximum in stress appearing at about 400° C. The different times at temperature before testing have had no noticeable effect. The effect of copper on the reduction in area is very marked, beginning at about 300° C and giving a minimum in reduction of area at about 600° C. A minimum is also obtained at about 200° C as is to be expected (Fig. 48).

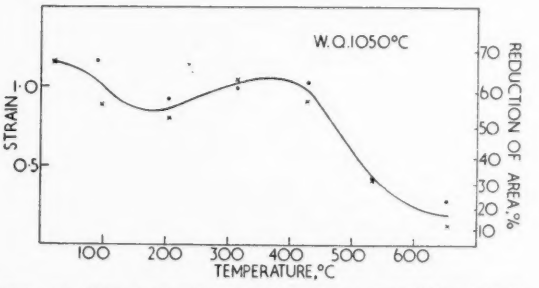

Fig. 45—Effect of 31% and 41% Ni on ductility at high temperature (steels 25 and 26)

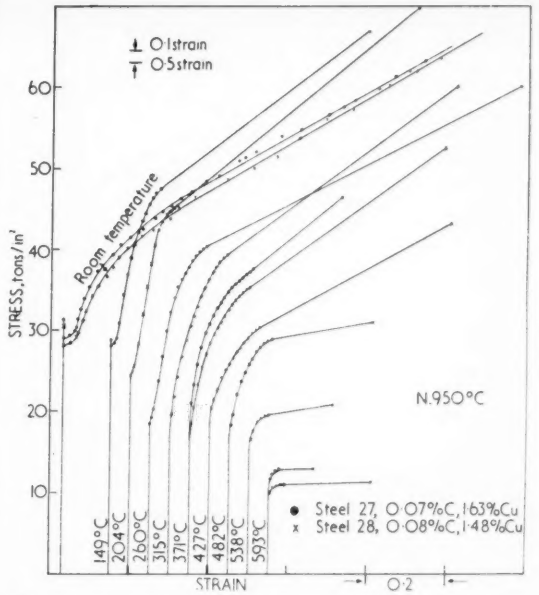

Fig. 46—True stress/strain curves for 1·5% Cu steels (27 and 28)

Precipitation of copper from steel on tempering has been clearly established, good evidence in the form of photomicrographs being given recently by Bain.[18]

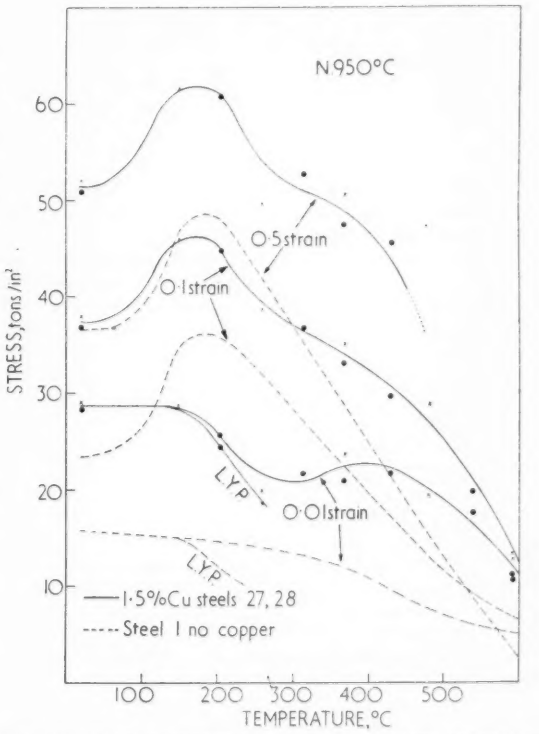

Fig. 47—Effect of copper on high-temperature strength

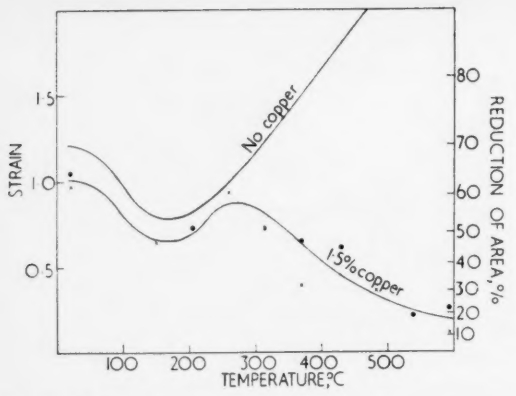

Fig. 48—Effect of copper on ductility at high temperature

Maximum hardening is obtained on tempering for 1 h at 500° C, but it would take a very long time at 400° C. Thus the maximum in stress at 400° C and 0·01 strain, and possibly even a lower temperature at higher strains, must surely result from more rapid precipitation of copper. Dislocations represent favourable sites for precipitation and it can thus occur at a lower temperature during straining than during tempering. A similar argument can be applied to some of the steels already tested. For example, with the 1·5% molybdenum steel, maximum hardening at room temperature, due to precipitation of molybdenum carbide on tempering, occurs after tempering for 10 h at 650° C. During straining of the normalized steel, however, maximum hardening occurs on straining at 500° C. It would appear, therefore, that precipitation of copper in the dislocations has a similar effect to precipitation of carbide.

REDUCTION OF AREA

As already noted, each maximum in stress results from the precipitation of a particular carbide in the dislocations formed during straining. But corresponding to each maximum in stress is a minimum in reduction of area. Thus it appears obvious that each minimum in reduction of area is also a consequence of precipitation. Before discussing this aspect any further, however, it is necessary to consider the mode of failure of the test pieces.

Microscopic examination of sections cut from test pieces broken at temperatures above about 300° C showed that failure was more or less intercrystalline in character. At a temperature approximating to that of a minimum in reduction of area (at least for all samples which showed a low reduction of area) the fracture was completely intercrystalline, and cracks could be detected at a considerable distance from the main fracture. Above this temperature the intercrystalline cracks did not begin to form until the test piece had necked down to some extent. The cracking and fracture were confined to a single plane and no cup and cone was formed. At a temperature below that of the minimum in reduction of area, the fracture appeared to be of the cup and cone type but it was observed that the flat area at the centre of the fracture showed intercrystalline cracks, whereas at the sides failure was by shear. In other words, failure was initiated by intercrystalline cracks near the centre of the test piece and changed to shear near the periphery.

Since the grain boundaries can be considered as an array of dislocations, it seems probable that precipitation occurs in these boundary dislocations during straining. The initiation of slip across the grain boundary will thus be impeded, and dislocations will tend to accumulate near the grain boundaries. As the boundaries become more rigid they will require to support a greater proportion of the stress, so that failure eventually occurs. Recovery or recrystallization will relieve the grain-boundary stresses and thus increase the reduction of area, as was the case with the low-titanium steel normalized at 950° C (Fig. 38). The interstitial and alloy elements have no great distance to travel to precipitate in ordinary dislocations, but have a considerable distance to go to reach the boundary dislocations. Thus the minimum in reduction of area should occur at a higher temperature than the corresponding maximum in stress. The experimental data show that such was the case.

At temperatures below 300° C, no evidence of intercrystalline failure could be detected in the fractured specimens, but at these temperatures the stress is high and one grain-boundary crack would create such a high stress concentration that general failure would result very quickly. It seems possible, therefore, that a simple mechanism would explain the minimum in reduction of area at 200° C.

The steels deoxidized with aluminium had in general a finer ferritic grain size than the steels without aluminium, and this may explain why the intensity of the minima in reduction of area is reduced by aluminium deoxidation. With a finer grain size there is a greater grain-boundary area and the effect of precipitation will be less pronounced. However, such an explanation does not appear to be complete, since the difference in structure was not very noticeable in the casts made in the small high-frequency furnace. Also, at temperatures away from the minimum, aluminium does not have much effect. In view of this difficulty the author would suggest that nitrogen is an important factor in determining the intensity of the minima in reduction of area. Admittedly it has no effect on the intensity of the maxima in stress due to different alloying elements, but in that case the rate of diffusion of nitrogen is not important. However, time is required to reach the boundaries, and since nitrogen is more mobile than carbon it may tend to accumulate in that region.

The vanadium steels showed a minimum in ductility which was not accompanied by a maximum in stress. In view of all the evidence on steels containing other alloying elements, there seems little doubt that precipitation of vanadium carbide is responsible for this minimum in ductility. A minimum in ductility is also obtained in the 30% and 40% nickel steels at high temperature with no maximum in stress. In such cases precipitation at the grain boundaries appears to be much more rapid than precipitation within the grains, so that failure occurs before the grains reach a high level of strength.

It is obvious from the above that grain-boundary failure is a complex problem. It is also one of great importance in practice. For the present purpose, however, it suffices to state that low ductility at high temperature appears to be related to precipitation at the grain boundaries. Further evidence dealing with the effect of combinations of alloying elements and with the effect of tempering before testing will be presented at a later date. A more detailed consideration will then be given to this problem.

SUMMARY AND CONCLUSIONS

By means of true stress/strain tensile tests carried out over a range of temperatures, the effect of individual alloying elements on the strength properties of low-carbon steel in the normalized condition has been studied. The following conclusions have been reached.

(1) In a simple steel containing a low alloy content one strain-age-hardening phenomenon was observed which gave a maximum in stress at about 200° C and a minimum in reduction of area at about the same temperature.

(2) The above phenomenon is related to the presence of interstitial carbon and nitrogen in steel. If nitrogen is removed by aluminium deoxidation the intensity of the phenomenon is reduced, since nitrogen augments the effect of carbon.

(3) It is postulated that iron carbide and nitride are precipitated in the dislocations during straining at a temperature where the interstitial solute has time to diffuse into the dislocations during the period of the test.

(4) Most of the hardening at about 200° C occurs in the initial stages of testing for the following reason. The supply of interstitial solute is limited so that only a certain number of dislocations can be filled with precipitate; when this number is exceeded, i.e. after a certain extension, the rate of work-hardening decreases abruptly.

(5) Before straining and at small strains there is sufficient carbon alone to fill the dislocations with precipitate. Thus, up to about 5% extension there is very little difference between the steels with and without aluminium deoxidation. At higher strains nitrogen is necessary to augment the supply of interstitial solute atoms and thus maintain the high rate of work-hardening. At room temperature little or no ageing occurs and there is thus little difference in the tensile curves of the two types of steel.

(6) The addition of a strong carbide-forming element modifies the strain-age-hardening phenomenon at 200° C and introduces a second strain-age-hardening phenomenon at a higher temperature.

(7) This second strain-age-hardening occurs at a higher temperature because it can only take place at a temperature where the alloying element can diffuse appreciably during the period of the test. Thus in this case the rate of diffusion of the alloying element is the controlling factor and not the rate of diffusion of carbon and nitrogen. In this temperature range precipitation of alloy carbide or nitride takes place in the dislocations formed during straining. A maximum in stress and a minimum in reduction of area is obtained, similar to that found at about 200° C.

(8) These secondary strain-age-hardening maxima in stress are not influenced by aluminium deoxidation, since the controlling factor is the rate of diffusion of the alloying element. There is ample carbon available at these temperatures to form alloy carbides and nitrogen is not necessary.

(9) Although the alloy carbide cannot precipitate at 200° C, the results at this temperature are influenced by the alloying element present in solution. Interstitial solubility is related to the affinity between the solute element and the solvent. The alloying element present has a greater affinity for carbon and nitrogen than iron, so that the interstitial elements tend to segregate into interstices adjacent to alloy atoms. If some or all of the interstitial solute atoms are fixed in such positions they are not free to migrate and precipitate in dislocations during straining. As a result the intensity of strain-ageing at 200° C is reduced by the presence of the alloying element in solution. Similarly, strain-age-hardening at room temperature is reduced.

(10) The extent to which an alloying element modifies strain-age-hardening at 200° C depends on at least three factors:

(i) The affinity between the interstitial solute and the alloying element.

(ii) The number of alloy atoms present, i.e. the atomic weight per cent in solution. If some of the alloying element is present in the carbide phase, or as oxide or nitride, it has no effect.

(iii) The number of interstitial solute atoms present. In this connection there is some evidence that molybdenum, vanadium, and titanium increase the solubility of carbon in steel. If the solubility is increased, this factor will tend to counteract the effect of factors (i) and (ii) above. The change in solubility of carbon and nitrogen in iron brought about by the addition of alloying elements is a subject on which little information is available. It would repay more detailed attention.

(11) The addition of sufficient vanadium or titanium completely suppresses strain-age-hardening at about 200° C since all the interstitial solute is firmly fixed at interstices adjacent to alloy atoms. However, at higher temperatures thermal agitation tends to free some of the interstitial solute. In these steels a maximum in stress was obtained at about 400–500° C which was attributed to the precipitation of manganese carbide. It is hoped to repeat these experiments on vanadium and titanium steels of low manganese content. If the maximum in stress disappears, it will prove whether or not manganese was responsible for the effects shown by the steels in the present series.

(12) Vanadium and titanium appear to be exceptions to the general rule in that no maximum in stress was observed. A minimum in reduction of area was, however, obtained in the vanadium steels, and an indication of one in the titanium steels which could be related to the presence of these elements. However, vanadium and titanium carbides form so slowly and at such high temperatures that their effect on strength was probably negligible in the testing time used. Recovery and recrystallization at these high temperatures would also influence the result obtained.

(13) Steels containing the non-carbide-forming elements nickel and silicon were investigated. The addition of these elements was found to have little or no effect on the strain-age-hardening characteristics of the steel.

(14) A 1·5% copper steel was also tested and the presence of copper was found to have a similar effect to the presence of a carbide-forming element.

(15) Two simple austenite steels containing nickel were tested and a strain-age-hardening maximum in stress and minimum in reduction of area were obtained at about 200° C. It appears that strain-age-hardening phenomena can occur in austenitic steel and are not confined to ferritic steels.

(16) The mode of failure at high temperatures was generally of the intercrystalline type, and it is probable that precipitation at the grain boundaries is responsible for this and for the minima in reduction of area. This subject is a complex one and further evidence is necessary before it will be possible to suggest a more definite mechanism of the mode of failure.

Acknowledgments

The author wishes to express his thanks to Mr. W. Barr, Director and Chief Metallurgist of Colvilles Ltd., for his continued encouragement and advice, and to him and his fellow Directors for permission to publish this paper.

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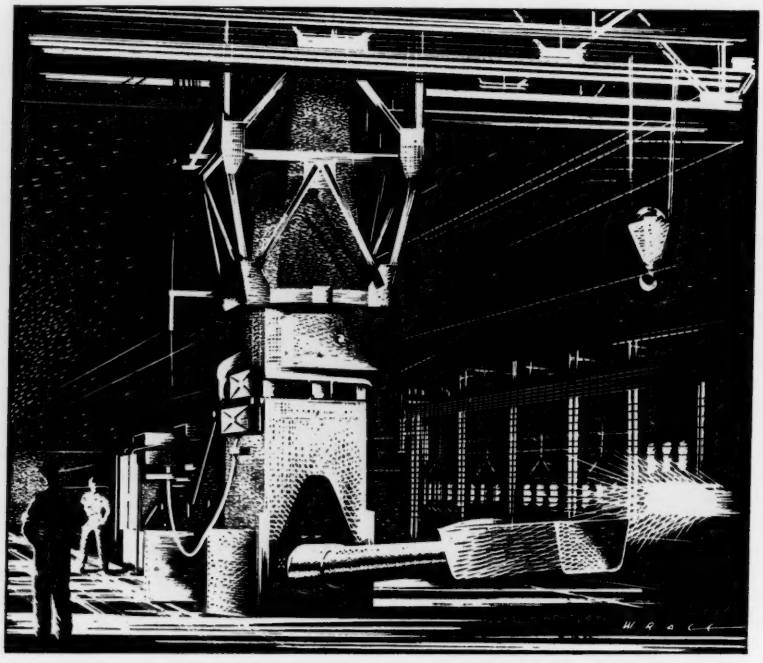

Correspondence on the Paper—

Reactions involved in the First-Stage Graphitization of Iron–Carbon–Silicon Alloys*

By W. S. Owen and J. Wilcock

Mr. **G. Östberg** (Stora Kopparberg Bergslags AB, Falun, Sweden) wrote: The factual material concerning the graphitization of Fe–C–Si–alloys presented in the paper by Dr. Owen and Dr. Wilcock, and in the preceding paper by Dr. Burke and Dr. Owen,[1] makes a valuable contribution to our knowledge of the graphitization process. The complicated nature of this process and the difficulty of formulating a theory which fits individual details is recognized by the authors when they state in the introduction that " it is necessary to examine the component reactions in detail " in order to arrive at something more than " an approximate description of the gross effect ". The present writer wishes to emphasize this by calling attention to some phenomena, which may affect the authors' interpretation of their observations. In cases where the authors' paper is based on the article by Burke and Owen, the latter will also be discussed in some detail.

First of all, some comment must be made on the use of the dilatometric method of investigation. The authors verify the assumption by Burke and Owen that the volume expansion of the iron is proportional to the increase in the volume of graphite. This does not agree with an earlier investigation on malleable iron by Smith and Palmer, which shows a discrepancy between the amount of graphite calculated from proportionality to expansion and the amount found by chemical analysis.[2] This point is important since dilatation has been and still is widely used as a measure of the amount of graphitization in various iron alloys. The authors' opinion about the general validity of the method would therefore be of great value.

Among the details of the graphitization the nucleation process deserves a particular analysis. The authors have adopted the opinion often held that graphite is nucleated at the interface between austenite and (ledeburitic) cementite. This idea, which is indeed attractive, needs, however, to be more precisely defined, before it can be used to explain the features of nucleation. Firstly, there is an austenite/cementite interface which is not always taken into consideration in these discussions, namely, the one of dissolving pro-pearlitic and perhaps also pearlitic cementite. Especially when the iron is heated up rapidly, there can be a considerable visible amount of such carbide left at the graphitization temperature. Cementite particles of this kind, scattered in the matrix between the areas of massive carbide, can be retained for an appreciable time, as can be seen for instance in Fig. 8 in the paper by Burke and Owen.†

From other fields of metallography (e.g. spheroidization of pearlitic cementite) there are indications that remnants of such carbides can persist on heating above their equilibrium range of existence, and then promote nucleation in further transformation.[3] Graphite nucleation at such particles seems to be difficult to corroborate by direct observation or to distinguish from the effect of other phase boundaries. Sulphides in malleable iron with a high sulphur/manganese ratio, for instance, have been shown to provide a substrate for graphite formation.[4] It should be remarked in this connection that Avrami's theory cannot be used as an argument in the discussion of this subject, until its postulate of a limited number of nucleation sites has been found valid. In the case of sulphide particles acting as nucleating agents, there is an increase in their number during the annealing by precipitation of such inclusions.

The process of nucleation has been studied by the authors in annealed specimens by observation of graphite particles, sometimes called ' nuclei ' by the authors, which have grown large enough to be visible in the microscope. It is worth mentioning here that graphite nuclei may well have been formed before the iron has arrived at the annealing temperature, an idea that has been extensively elaborated by Saltykov.[5] The authors' graph of the number of graphite nests‡ per unit volume plotted against time shows a linear increase up to a maximum. However, if the change in number of graphite nests per unit section surface of the specimens is followed with closer intervals of time, as has been done by the present writer, an interruption in the linearity will be found. In an early stage, when small particles of non-eutectic cementite still can be seen, the time function of graphite nests passes a maximum and a minimum and then resumes the increase up to the ultimate maximum. A transformation of the function into unit volume would not change this trend. There are indications of such a phenomenon also in the curves of Brown and Hawkes.[6] The phenomenon mentioned indicates that the sequence of reactions is more complex than was assumed by the authors. The validity of the assumption of a constant nucleation frequency as a basis for a hypothesis for the overall reaction is thereby seriously questioned.

* *J. Iron Steel Inst.*, 1956, vol. 182, pp. 38–43.

† The amount of graphite in the specimen shown in Fig. 8 is quite considerable. It might be thought that this specimen should then have passed the point that the authors define as the ' start of reaction ' or the end of the ' incubation period '. However, according to the data given, this is not the case, as can be deduced from Fig. 2, showing the ' start ' point for an iron with a higher silicon content, and hence a shorter graphitization time. The definition and the determination of the ' start of reaction ' and the ' incubation period ' therefore seems to have little significance from a metallographic viewpoint. Furthermore, with regard to the variation of the number of graphite nests, the ' start ' point does not rationally conform to the supposition that during the time of reaction that follows the ' nucleation frequency ' should be constant. This becomes clear when comparing Figs. 3 and 11: at the point defined as the ' start of reaction ' the steady increase in number of nests already begins to fall off.

‡ The term ' flake nest ' or ' graphite nest ' is here preferred to ' nodule ', because the latter term is often also used for spherulite. The lack of distinction between these different forms of graphite crystal aggregate seems to have misled investigators in this field. Burke and Owen use the name ' spherulite ' for flake nests which are admitted to be not even " as perfect in form as those usually found in comparable commercial alloys".

Another feature of the 'nodule count', which the authors discuss at some length, is the drop in number of graphite nests at a late stage of the annealing. This phenomenon is attributed to differences in the radii of the graphite nests. This would be correct if the radius referred to a (single) crystal with a smooth surface. However, the graphite nests consists of flakelike units, pointing outwards, seemingly without any preferred direction. The surface of each of these flakes displays a variation of radius of curvature from almost zero at the edge of the flake to nearly infinity at the sides. It is thus not very meaningful to talk about the radius of curvature of a graphite nest. If the authors' reasoning on this point is to have any possible meaning, there has to be a relation between the mean diameter of the nest and the size of the flakes, which remains to be proved. A microscopical study reveals another feature of the structure of the graphite nests, at least in those formed in irons with a high S : Mn ratio and particularly at a high temperature, namely their growth by branching.[4] It is felt that this process is an important one which is worth closer attention.

Among the factors that somewhat disturb the picture of growing graphite and dissolving cementite is the uneven distribution of silicon in the structure of the iron, owing to the fact that the cementite does not dissolve appreciable amounts of Si, according to one of the authors[7] supported by others.[8] New austenite, formed by transformation of cementite during graphitization, inherits a low Si content. As the silicon content of the austenite is levelled out by diffusion to areas of former cementite from the pre-existing bulk, there is a corresponding change in carbon solubility which is a function of the silicon content. In regions where the silicon content is lowered, graphite will then be retarded in its growth or perhaps even dissolve and vice versa. This effect is superimposed upon the basic growth process. This redistribution of silicon may play a part in the dissolution at a late stage of some graphite particles, as discussed above. Parallel to the effect of the transformation of cementite on the silicon content is the enrichment of the silicon in front of the advancing growth front of graphite in austenite, mentioned by Burke and Owen. This enrichment is thought by the present writer to have some bearing upon the well known fact that cementite particles remote from growing graphite start to dissolve before those adjacent to it have completely disappeared. The effect is probably due to an influence of silicon on the gradient governing the carbon diffusion.

In recent years the mechanical problems associated with the growth of graphite have aroused some interest. Self-diffusion of iron mentioned by Burke and Owen has early been pointed out as a requirement for the growth of graphite by Russian workers, notably Bunin and his co-workers.[9] Creep in the matrix is another mechanism which has been suggested by Hillert[10] in the case of graphite growth in ferrite. It is felt that such processes, although not yet sufficiently explored to be applied in a discussion of malleable iron, could well be of importance in determining the rate of graphitization.

However, it is not enough to consider the possible rate-determining reactions separately without regard to the structural interrelation between the phases involved. It may well happen that the rate-determining role changes during the annealing time from one component reaction to another, depending upon the structural changes created by the graphitization. It may also be that the growth of different graphite nests at a certain time can be governed by different reactions. Therefore, it seems as if the dilatometric method, which registers at a certain moment the net effect of reactions at various stages, is less suited to give an account of the complicated mechanism of graphitization in alloys of the malleable iron type.

AUTHOR'S REPLY

Dr. **J. Wilcock** (University of Birmingham) and Dr. **W. S. Owen** (Massachusetts Institute of Technology) wrote: We endorse Mr. Östberg's general thesis to the effect that close attention to detail is necessary when considering the graphitization of iron–carbon–silicon alloys. However, we suspect that many of the interesting points which he raises are based upon observations of alloys containing manganese and sulphur. To transfer these thoughts to the pure ternary alloys may be misleading.

In addition to the present work of collating dilatometric and microscopic data, it has previously been shown by Brown and Hawkes[6] that a master reaction curve was obtainable using data derived from different methods of investigation. The fact that such graphs could be superimposed is a strong argument against the conclusions of Smith and Palmer.[2] The linear expansion most certainly cannot be exactly equal to the cube root of the volume increase of graphite since there is a volume change due to the solution of cementite. The latter effect is very small compared with the former and we think that the data represented in Fig. 5 of our paper demonstrate this fact.

We cannot agree with Mr. Östberg's statement that "the [dilatometric] determination of the start of the reaction seems to have little significance from a metallographic viewpoint". It is true that in the earlier determination by Burke and Owen[1] the coincidence of dilatometric and metallographic observations was not established. The growth curves did not extrapolate to zero reaction time (in some cases the intercept on the time abcissa was negative even when extrapolated to zero heating time). As discussed in the paper, this was evidently due to appreciable errors in the measurement of the selected spherulite diameters when the number of spherulites was small. No graphite was found in specimens annealed to zero reaction time. (Unfortunately, the experimental work was performed more than five years ago and we have been unable to trace the origin of the misleading information attached to Fig. 8.) The matter was re-examined subsequently and more recent data, obtained by much improved methods of lineal analysis, are reported in the paper at present under discussion. It is now seen that the curves do extrapolate reasonably well to the origin representing the start of the reaction as shown by the dilation curves. The nucleation frequency curves, if extrapolated linearly, intercept the abscissa at positive reaction time. Thus, there appears to be an initial period during which the number of spherulites per unit volume increases only very slowly with time. The formation of submicroscopic graphite nuclei at the interface of dissolving proeutectoid cementite during heating may be important during this early period of the reaction, and it is interesting that Mr. Östberg has determined nucleation frequency curves which reflect this phenomenon. However, at the stage when the graphite spherulites were clearly defined in our alloys almost all of them were located at an interface between massive carbide particles and austenite. When measuring the total cementite/austenite interfacial area we included all the cementite which could be seen under the microscope and did not attempt to differentiate between particles on the basis of their probable origin. The alloys did not contain sulphur or manganese and thus Mr. Östberg's comments based upon observed effects in the presence of these elements do not seem to be relevant. A study of graphitization in alloys

containing sulphur and manganese has recently been completed at Liverpool and we will consider the points which have been made when we report on this work.

The graphite in our high-purity iron–carbon–silicon alloys cannot be described as in the form of ' flake-nests '. The nodules appear as imperfectly formed spheres, i.e. spherulites. Of course, they are not single crystals, but this does not invalidate the use of the concept of solution potential as a function of the radius. We agree that the re-solution of graphite flake-nests in alloys containing manganese and sulphur (if, in fact, such re-solution occurs) will require a more elaborate explanation.

We are in complete agreement with the contents of the last three paragraphs of Mr. Östberg's discussion. In our paper we attempted to emphasize the importance of silicon partition and diffusion in the graphitization process.

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9 (*a*) K. P. Bunin and N. M. Danilchenko: *Doklady Akademii Nauk SSSR*, 1950, vol. 72, pp. 889–890.
(*b*) K. P. Bunin and G. I. Ivantsov: *Ibid.*, pp. 1051–1053.
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Correspondence on the Paper—

The Solubility of Nitrogen and Formation of Silicon Nitride in Iron–Silicon Alloys*

By E. T. Turkdogan and S. Ignatowicz

Mr. **Alan U. Seybolt** (General Electric Company Research Laboratory, Schenectady, U.S.A.) wrote: The authors appear to find some anomalies which suggest lack of true equilibrium. For example, take the $(p_{NH_3}/p_{H_2}^{\frac{3}{2}})$ versus $\%N_2$ plots in Fig. 1. The ordinate here is proportional to the nitrogen activity or nitrogen potential. For any silicon content, the percent nitrogen in solution rises until a critical nitrogen potential is reached: this critical potential is that required to form Si_3N_4. If one compares the potentials for 0·9% Si at 500° and at 600° C, it is observed that the figures are approximately 0·037 and 0·018, respectively. Similarly, at 1·26% Si, the potential at 500° C is considerably larger than at 600° C. This behaviour is anomalous because one expects that at higher temperatures the nitride becomes less stable, hence requiring a higher nitrogen pressure to allow it to exist.

In Fig. 4 the data for the 0·90% Si alloy appear to exhibit a monatonic relationship between nitrogen potential and nitride formation. However, in the richer alloys, it looks as if at a potential of approximately 0·05 the nitrogen potential had reached a sufficiently high level to convert all the remaining silicon in solution to Si_3N_4. Drawing a vertical line between E and F, and between G and H, attains a nitrogen content corresponding to complete conversion. From the stoichiometry of Si_3N_4, it is found that $\%N = (0·667)\%$ Si. For the three alloys this relationship predicts 0·6%, 0·84%, and 1·89% N, values which are quite accurately obtained in Fig. 4. I believe that the evidence here is that, with increasing silicon content, the nitrogen potential required to form Si_3N_4 drops sharply. This drop is probably so drastic that with silicon contents of over 1%, it becomes exceedingly difficult to control the $p_{NH_3}/p_{H_2}^{\frac{3}{2}}$ ratio at the equilibrium value. The same thought was expressed by the authors near the bottom of p. 201, and was given as the reason for going to nitrogen–hydrogen mixtures. This is another reason for expecting that the latter part of the higher silicon curves of Fig. 4 may not represent equilibrium, as suggested above.

At the end of the paper, some calculations of displacement of curves are made. Bearing in mind the possibility of lack of equilibrium in Fig. 4, it is perhaps unjustifiable to draw any conclusions based on the shape of the curves shown there. On p. 206 the thought is expressed that because of a difference in shape in the curves of Fig. 4, lack of equilibrium might have occurred in the 0·9% Si alloy. Here again, it might be repeated that the evidence seems to point to lack of equilibrium in the richer alloys. Much more convincing evidence of a true equilibrium condition could have been obtained by changing the gas ratio for a given composition and temperature, and observing corresponding losses or gains in nitrogen content, i.e. approaching equilibrium from both sides.

The authors' comments about the difficulty of nitriding to Si_3N_4 under various conditions fits in with my own experience. I have just completed a similar study to that being discussed, except that silicon alloys up to 12% Si were investigated. Also, while nitriding techniques were used to form some Si_3N_4, equilibrium was obtained by use of nitrogen–argon mixtures containing a little hydrogen. In these experiments it was found that silicon nitride was not formed by nitrogen gas, even in the presence of about 1% hydrogen. Rather, the presence of ammonia was necessary. However, some nitride having once been formed, it was then possible to secure equilibrium with nitrogen gas. The reason for this is apparently due to some catalytic effect rather than to the thermodynamics of the situation. In addition, it was found that the rate of formation of Si_3N_4 was greatly dependent

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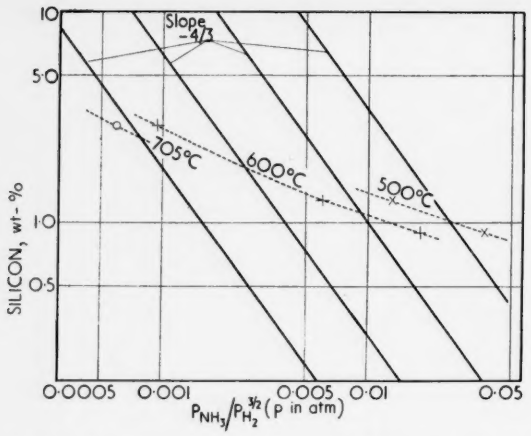

Fig. A—Log [Si] versus log ($p_{NH_3}/p_{H_2}^3$) for the authors' results

upon the moisture content of the gas and upon the nature of the metal surface. A freshly abraded metal surface together with a slightly moist gas (perhaps a dewpoint of around 0° C) resulted in a rate of Si_3N_4 formation at 700° C which was so fast it was difficult to prevent complete conversion to nitride; 4% ammonia in hydrogen was used as the nitriding gas. On the other hand, if the gas was dried at — 50° C, no appreciable nitriding would be noted after many hours.

Mr. **D. I. Cameron** and Mr. **A. Oates** (Lysaghts Works Pty., Ltd., Newcastle, Australia) wrote: The extremely difficult experimental work undertaken by the authors has provided results of considerable practical interest. They have established a general pattern of behaviour, but there are several points which seem to us to need clarification.

We have found, along with others[1] the greatest difficulty in avoiding oxidation in silicon steels containing 1% or more silicon. This shows up as a surface tarnish even in carefully dried atmospheres of 100% hydrogen. In 4% silicon steels this tarnish appears in 100% hydrogen at all temperatures below about 1300° C, even with dewpoints as low as — 40° or — 50° C. At lower temperatures the dewpoint requirements are even more difficult. If a treatment is given in 100% hydrogen at 600° C to any silicon alloy containing more than 1% silicon, the dewpoint would have to be less than — 100° C in order to avoid all possibility of oxidation. It is known that oxidation of silicon steels interferes with nitriding and it has not been made clear by the authors that the efforts to avoid oxidation have been successful. We would, therefore, like to know whether the authors did in fact manage to achieve a complete absence of tarnish on their samples and whether any attempt was made to determine if any silica was formed by oxidation of the samples during the anneals.

It appears quite definite that the silicon nitride that occurs in the presence of silicon in steel up to 3% is Si_3N_4 or something very close to it in composition. The possibilities of a different formula or of a mixed iron silicon nitride are ruled out, notably by the presence of the intercepts which the authors show in Fig. 4: If we assume that the nitride is Si_3N_4, or something very close to it, then for equilibrium with the pure nitride

$$\log K = 3 \log a_{Si} + 4 \log a_N$$

On the assumption that silicon obeys Henry's law, for a given temperature we may write

$$\log K' = 3 \log [Si] + 4 \log (p_{NH_3}/p_{H_2}^3)$$

Thus a plot of log [Si] versus log ($p_{HN_3}/p_{H_2}^3$) should have a slope of — 4/3. In Fig. *A* the authors' results from Table III have been plotted, the nitrogen pressure values for the 2·83% Si alloys having been converted to ($p_{NH_3}/p_{H_2}^3$) ratios by using Turkdogan and Corney's recommended values for the free energy of formation of ammonia.[2]

From Fig. *A* it may be seen that the authors' results lie in curves having slopes far from — 4/3; instead they lie on a slope of about — 1/3. This strongly suggests that true equilibrium has not been reached, since there is no suggestion of any nitride of composition approaching Si_3N as these results imply. A similar conclusion is reached if the estimated equilibrium nitrogen contents of the steel are plotted against the silicon content.

The authors have not in any of this work approached equilibrium from the nitrogen-rich side, though it should be readily possible for the high-silicon alloys, even in simple nitrogen atmospheres. The two-way approach to equilibrium is particularly important in dealing with surface reactions of sheet material. As Darken[3] has shown, the process is controlled by the relative abundances and diffusion rates of the elements concerned. Persistent steady states may be set up in the surface layers short of the final equilibrium. With silicon contents of 1–2% these layers could be quite thin compared with the 0·002 in. thickness of the material. It would be valuable to know whether micro-examinations were carried out and what the distribution of the nitride was throughout the thickness of the material.

As the authors have pointed out, the displacement of the curves in the upper part of Fig. 4 indicates that the 0·9% silicon, at least, has not reached equilibrium. The authors' Table IV only goes down to 0·02 for the gas ratio, and if the same treatment is extended to values below this it will be found that not even the 1·26% or 2·83% silicon steels have reached equilibrium at these lower nitrogen potentials.

AUTHORS' REPLY

Dr. **E. T. Turkdogan** and Mr. **S. Ignatowicz** (B.I.S.R.A.) wrote in reply: In all the experiments, with 0·90%, 1·26%, and 2·83% silicon iron alloys, the brightness of the sample was not altered after heating in pure hydrogen, nitrogen–hydrogen, or ammonia–hydrogen mixtures, at or above 500° C. In these experiments ammonia was dried by passage over potassium hydroxide pellets, and nitrogen and hydrogen dried by passage through vertical columns (2 ft high × 1 in. dia.) containing magnesium perchlorate. The water vapour partial pressure of the gas dried in this way is of the order of $2·5 \times 10^{-6}$ atm. However, simple cold traps (at about — 50° C) are not as efficient in drying gas streams as long columns of anhydrone.

The effect of moisture on the rate of nitride formation in silicon steel, mentioned by Seybolt, is of great interest. The authors, however, have tried wet nitrogen as a nitriding atmosphere.

The argument put forward by Cameron and Oates with regard to the relationship in Fig. *A* cannot be accepted. There is no justification in their assumption that "a solution of silicon in solid iron obeys Henry's law." On the contrary, the only deduction that can be drawn from the relationship in Fig. *A* is that there is a distinct departure from Henry's law.

The nitrogen contents of the original alloys containing 0·90% and 1·26% Si were 0·013% and 0·012% respectively. In some of the experiments the samples were treated initially in hydrogen to reduce the nitrogen content of the sample to about 0·0003%. A number of experiments, however, were carried out without the initial denitriding treatment. Unfortunately, this point was not made clear in the paper. The arrows in Fig. *B*, reproduced from the paper, indicate the direction of approach to equilibrium. In view of this evidence, there remains no doubt about the state of equilibrium in the authors' experiments. It is, however, regretted that these data were not included in the paper.

A number of nitrided samples were examined under a microscope indicating that nitriding was uniform. This is illustrated by a photomicrograph (Fig. *C*), which is the cross-section of sample 2·83% silicon–iron nitrided in 10% ammonia + 90% hydrogen at 600° C for one week.

In view of the above evidence, the statements made by Cameron and Oates in the last two paragraphs of their contribution are not valid.

The statements made by Seybolt about the relationship in Fig. 4 of the paper are confusing. In order to clarify Seybolt's interpretation of this diagram the following statements may be qoted from the original paper. " From thermodynamic considerations it may be stated that for a given temperature and gas ratio $p_{NH_3}/(p_{H_2})^{\frac{3}{2}}$ the concentration of silicon in solution will be the same, irrespective of the total silicon content of the alloy and the composition of the nitride. In other words, for any curve in Fig. 4, a given gas ratio will correspond to a certain silicon concentration in solution, provided the activity of the silicon nitride is not affected by the total silicon content of the alloy." Another quotation is as follows: " It is reasonable to assume that, at gas ratios

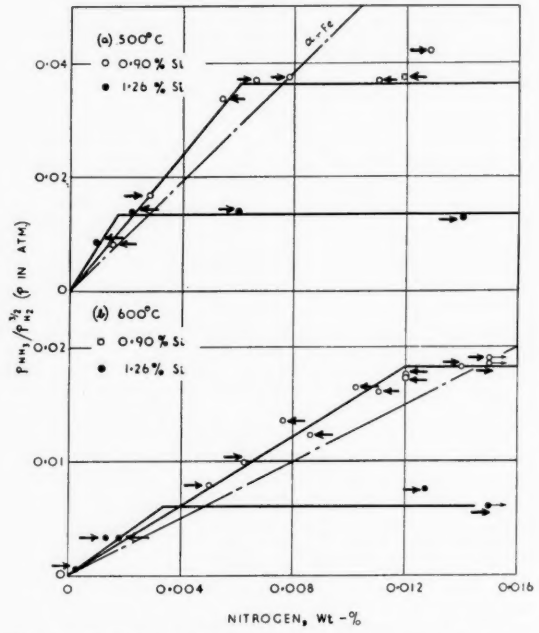

Fig. B—Solubility of nitrogen and formation of a silicon nitride in α-iron–silicon alloys at 500° and 600°C. Arrows indicate the direction of approach to equilibrium

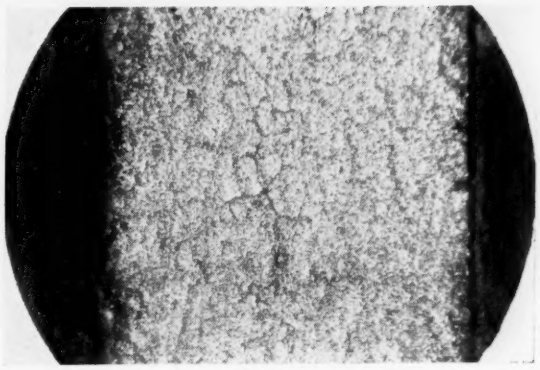

Fig. C—Cross-section of 2·83% Si alloy nitrided in 10% ammonia + 90% hydrogen for one week at 600° C × 350

very close to that for the formation of Fe_4N, almost all the silicon will be converted to silicon nitride, and the total nitrogen in the alloy will consist of the nitrogen required by silicon to form a nitride, plus nitrogen in solid solution in iron (0·1% nitrogen at 600° C in equilibrium with Fe_4N). The arrows marked on the horizontal broken line in Fig. 4 correspond to complete conversion of silicon to Si_3N_4. It is interesting to note that the extrapolations of the curves to the Fe_4N plateau agree well with the positions of the arrows."

It is clear from the above statements that the shape of the curves in Fig. 4 depends on (i) composition and activity of the silicon nitride, (ii) activity of the silicon in solid solution, and (iii) equilibrium constant of the reaction between silicon in solution, ammonia, silicon nitride, and hydrogen. Since there are no data to evaluate any of these three facts, nothing definite can be said about the shape of the curve for 0·90% silicon alloy in Fig. 4.

Since the publication of this paper some work has been carried out on the properties of pure silicon nitrides. A brief statement made in the paper can now be verified. That is, there appear to be two nitrides of silicon having almost the same compositions, i.e. corresponding to Si_3N_4, but with different crystal structures. The X-ray diffraction patterns and compositions of the nitrides extracted from iron–silicon–nitrogen alloys are in complete agreement with those of α-Si_3N_4. When the nitrided alloys are subjected to X-ray examination, prior to any chemical treatment, the nitride diffraction lines obtained are different from those of α-Si_3N_4 and β-Si_3N_4. This suggests that the silicon nitride in the alloy might be in the form of a complex nitride which decomposed to give α-Si_3N_4 during the chemical extraction process. Although the authors have no data on the activity of silicon nitride in iron alloys, the above statements, based on the results of recent work, may imply that the activity of silicon nitride in the alloy might vary with (i) the amount of the nitride, (ii) nitrogen potential of the system, and (iii) temperature.

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2. N. S. Corney and E. T. Turkdogan: *J. Iron Steel Inst.*, 1955, vol. 180, pp. 344–348.
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The Physical Metallurgy of Low-carbon, Low-alloy Steels containing Boron

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SYNOPSIS

The physical metallurgy of low-carbon, molybdenum–boron steels has been studied and a theory for the action of boron is suggested. The main effect of boron in steel of this composition is to increase the period of induction for polygonal ferrite formation and this effect increases linearly with boron content up to 0·002%. Polygonal ferrite formation is retarded to a much greater extent than bainitic ferrite and consequently a marked bay is produced in the isothermal transformation diagram, which allows bainitic ferrite to form over a wide range of cooling rates.

Boron is only effective at air-cooling rates in the presence of molybdenum, and it has been shown that this element increases the period of induction for polygonal ferrite up to at least 0·5%. Carbon has no appreciable effect on the transformation diagram over the range 0·01–0·20% C.

It is suggested that the enhanced mechanical properties are due to the very small sub-grain size of the bainitic ferrite which results from the low transformation temperature. The tempering characteristics of this type of steel have been studied and it has been shown that these steels are very resistant to tempering. Final softening occurs fairly rapidly and is associated with grain growth of the bainitic ferrite.

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THE DEVELOPMENT of a low-carbon, molybdenum steel containing boron has been described by Bardgett and Reeve[1] who showed that such a steel has a tensile strength of 40 tons/in^2 with an accompanying yield strength of 30 tons/in^2, and these properties can be obtained with a wide range of section sizes in the A.C. condition. This paper considers the physical metallurgy of these low-carbon, molybdenum–boron steels, and a theory for the action of boron is suggested which can also be applied to the more general question of the effect of boron in increasing the hardenability of low-alloy steels.

It is a well-established fact that boron retards austenite transformation reactions which nucleate predominantly at grain boundaries. It is also known, however, that the effect of boron decreases as the carbon content of the steel increases, until it is completely ineffective at about eutectoid composition. This leads to the conclusion that boron does not retard the formation of pearlite. Consequently, it appears that the main effect of boron is to retard the nucleation of pro-eutectoid ferrite at austenite grain boundaries and, to a lesser extent, the upper bainite reaction which is nucleated by ferrite.[2] The effect of boron is therefore restricted to reactions which occur mainly above the nose of the isothermal transformation diagram.

Hardenability is largely controlled by the period of induction at the nose of the isothermal transformation diagram and the effect of boron can be assessed by a consideration of a number of typical transformation diagrams (Figs. 1*a–e*). When the nose of the isothermal diagram is controlled by the '*C*' curve of pro-eutectoid ferrite (Fig. 1*a*) then boron, by retarding this reaction, will increase martensitic hardenability. On the other hand, if the pearlite reaction controls the nose of the transforma-

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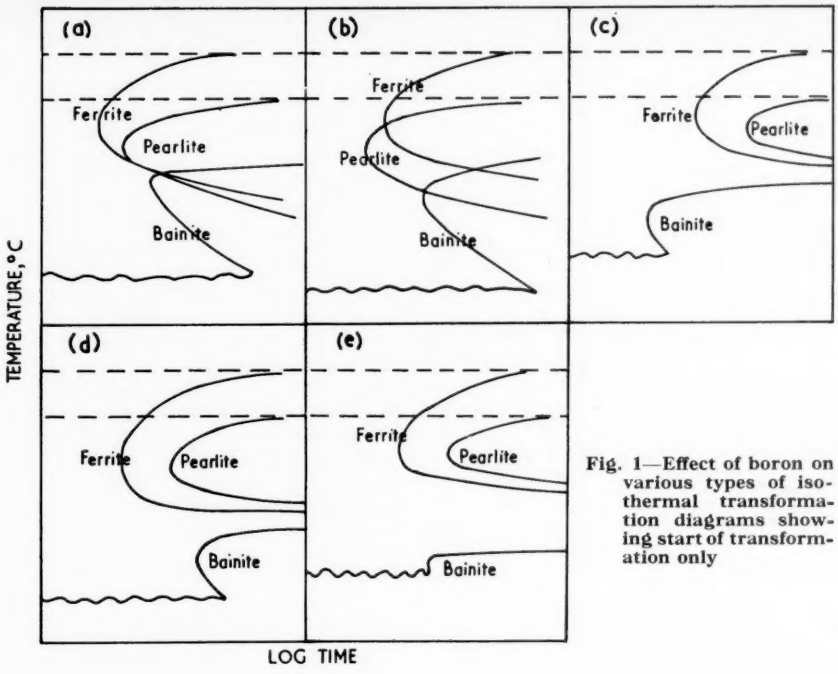


Fig. 1—Effect of boron on various types of isothermal transformation diagrams showing start of transformation only

tion diagram (Fig. 1*b*), boron will not increase the martensitic hardenability, but may cause a slight increase in bainitic hardenability. In diagrams where the ferrite and pearlite '*C*' curves are separated from the bainite '*C*' curves by a relatively stable austenitic region (Figs. 1*c–e*) the relative positions of the '*C*' curves will determine the effect of boron. In Fig. 1*c*, since the low-temperature bainite reactions would be unaffected, the addition of boron, by retarding pro-eutectoid ferrite, would increase bainitic hardenability but leave the martensitic hardenability unaffected. If the diagram was the unusual type shown in Fig. 1*d* with the ferrite '*C*' curve to the left of the bainite '*C*' curve, then the addition of boron could produce an increase in martensitic hardenability. Finally, if the diagram was of the type shown in Fig. 1*e*, with the bainite region very close to the M_s line, the bainite would have a structure and properties very similar to martensite, and a boron addition could be regarded as increasing the martensitic hardenability.

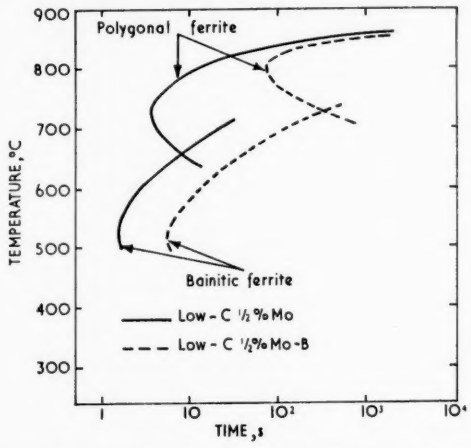


Fig. 2.—Isothermal transformation diagrams for low-C, ½% Mo steels with and without boron (showing start of transformation only)

The mechanism by which boron increases hardenability has attracted considerable interest, and it is now generally agreed that it does so by its ability to segregate or adsorb to austenite grain boundaries. The boron atom is small, belonging to the series carbon, nitrogen, boron, and hydrogen, and like these other elements it enters into interstitial solid solution in austenite. The atomic diameter of boron is 0·95Å compared with 0·75Å for carbon, so that to reduce the strain energy of the lattice, boron will diffuse to the largest interstices which will normally be found at the grain boundaries, and consequently will reduce the grain boundary surface energy. Similar processes have been described for carbon[3] and nitrogen.[4] Confirmation for this is provided by the following observations:

(i) Boron retards grain-boundary nucleated ferrite reactions

(ii) Boron forms a boron-rich phase at the austenite grain boundaries when the solubility is exceeded

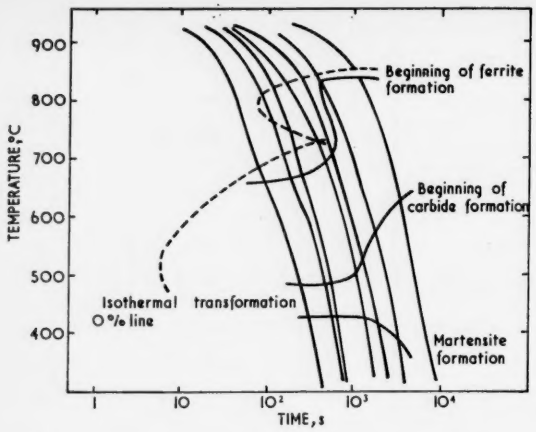


Fig. 3—Continuous cooling and isothermal transformation diagrams for low-C, ½% Mo steels

(iii) Electron microscopical examination of boron steels shows that the grain boundaries stand up in relief on the surface of the specimen.

The migration of an interstitial element to a grain boundary will depend to some extent on temperature. At temperatures below about 400° C, the rate of diffusion may be so slow that grain boundary segregates do not form. On the other hand, at temperatures as high as 1200–1300° C, the thermal activation may be high enough to disperse the grain boundary ' atmosphere ' which may account for the reported loss of hardenability when boron steels are austenitized at high temperatures.[2, 5] However, at favourable temperatures migration to grain boundaries will occur rapidly, although adsorption is not likely to be 100%.

Migration to form a grain boundary ' atmosphere ' means that the analysed boron content is only an average one and bears little relationship to the boron concentration at the grain boundary which is producing the increase in hardenability. In fact the effectiveness of boron will depend upon grain size and it has been observed in experiments that a boron content which is too low to produce a marked increase in hardenability can be made effective by the use of a higher austenitizing temperature. This may well be the reason why such varied boron contents have been reported in the literature as being the critical amount for improved hardenability.

In this paper, attention is confined to bainitic structures and the associated hardenability of low-carbon, low-alloy steels containing boron. The isothermal transformation diagrams of the base type of steel, with and without boron, are shown in Fig. 2. The main effect of boron is to retard the formation of high-temperature or polygonal ferrite, and a lesser effect is to retard the formation of bainitic ferrite. The result of this is that bainitic ferrite can be formed from the austenite at air-cooling rates, and it is the substitution of bainitic ferrite for polygonal ferrite which produces the improved mechanical properties in the steel.

CONTINUOUS COOLING TRANSFORMATIONS OF LOW-CARBON, MOLYBDENUM–BORON STEELS

The main feature of these steels is that they form bainitic ferrite during air cooling. It will be appreciated from Fig. 2 that at very slow cooling rates a soft polygonal ferrite (hardness about 160–180 D.P.N.) will form, and that as the rate of cooling is increased there will be a sudden change to a harder bainitic ferrite (hardness about 200–220 D.P.N.). A more accurate representation of this change is given in Fig. 3 which is a continuous cooling transformation diagram for a low-carbon, molybdenum–boron steel.

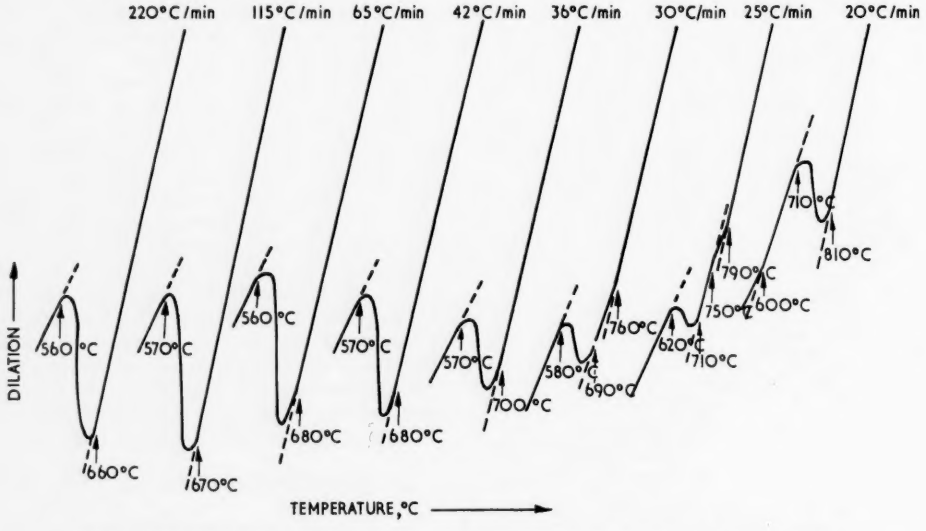


Fig. 4—Effect of rate of cooling on transformation of low-C, ½% Mo–B steels

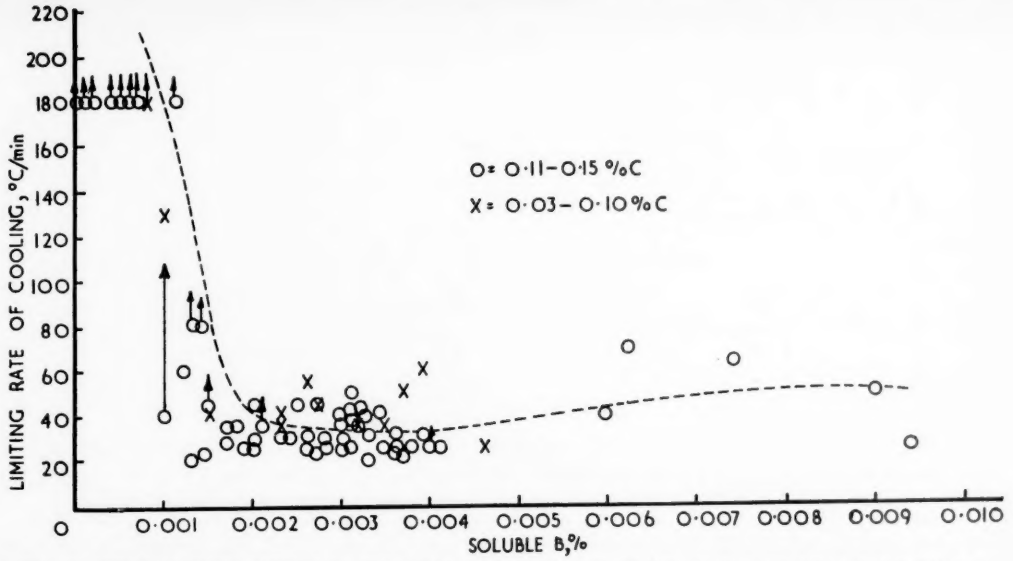


Fig. 5—Effect of boron content on the L.R.C. value of low-C, Mo–B steels containing 0·4–0·5% Mo

The isothermal transformation diagram is also drawn in Fig. 3 to show the similarity between the two types of diagram.

The slowest rate of cooling (in °C/min) at which no polygonal ferrite is formed has been called the *Limiting Rate of Cooling** or L.R.C.[1] Cooling rates cannot properly be compared if taken over actual transformation ranges and therefore the rate through the temperature range 900–870° C has been arbitrarily adopted for comparisons. These rates of cooling can readily be correlated with the rates of free-air cooling at the centres of bars or plates, and typical comparisons are shown in Table I.

The L.R.C. is determined dilatometrically and a typical series of dilation curves for various cooling rates is shown in Fig. 4. In this example, the L.R.C. is 30° C/min and the values quoted in Table I show that the desired bainitic ferrite structure will be produced in A.C. sections of commercially useful sizes. In fact the L.R.C. test is really a hardenability test for use with air-hardening materials forming low-carbon bainites, and the following paragraphs describe the effect of the main alloying elements on this bainitic hardenability.

* The Limiting Rate of Cooling should not be confused with the critical cooling velocity which implies the complete suppression of all non-martensite products of transformation.

Table I

COMPARISON OF COOLING RATES

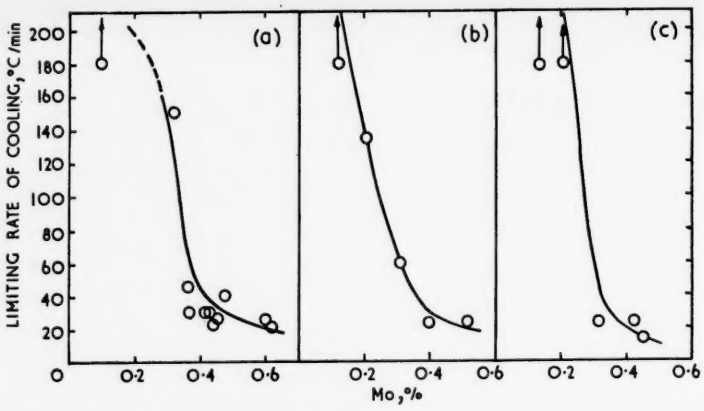
L.R.C. Value, °C/min	Corresponding to rate of air cooling at centre of	
	Bar, in.	Plate, in.
150	1	0·5
75	2	1
50	3	1·5
38	4	2
25	6	3

Effect of Different Elements on L.R.C. Value

(*a*) *Boron*

All the experimental results are shown graphically because of the large amount of numerical data involved. The effects of soluble boron content on the L.R.C. value is shown in Fig. 5. As measurements of L.R.C. values were not made above 180° C/min, the progressive change of L.R.C. with boron content at low concentration is not revealed as clearly as it is by the later isothermal transformation data. Nevertheless it is evident that as the boron content increases, the L.R.C. rapidly decreases to a general value of 20–30° C/min. When taken in conjunction with the isothermal transformation data (*see* Fig. 12), these results confirm the progressive effect of boron, which has been reported by several other workers[2, 6, 7] and is in contradiction to some of the previous literature which indicated that a critical boron content was necessary before the increase in hardenability could be detected.[8] In these steels, increasing boron causes an increase in hardenability up to 0·002% soluble boron,† but further increases in the boron content have no greater effect.

† Throughout this paper, boron figures quoted are acid soluble boron content.



(*a*) ½% Mn–B (*b*) 1% Mn–B (*c*) 1½% Mn–B

Fig. 6—Effect of Mo on L.R.C. of low-C, boron steels at varying Mn contents

(b) Carbon

In plain carbon steels, increasing carbon produces a marked decrease in the critical cooling velocity.[9] A similar effect might have been expected on the L.R.C. value, but in ½% Mo steels (in the presence of 0·002% B) increasing carbon from 0·01% to 0·15% had no appreciable effect on the L.R.C. (*see* Fig. 5). This effect is shown very clearly during the consideration of the isothermal transformation characteristics.

(c) Molybdenum

As with boron, because L.R.C. measurements were not made above 180° C/min, the variation of L.R.C. at low molybdenum contents cannot be accurately assessed. Nevertheless, Fig. 6 shows that with manganese contents up to 1·5%, an increase in molybdenum up to 0·4% causes a marked increase in hardenability, but with greater molybdenum contents the rate of increase in hardenability slows considerably.

(d) Manganese

From the information given in Fig. 6 molybdenum has the same effect at three different manganese levels and so it must be concluded that manganese itself does not have much effect on the L.R.C. This might be expected from manganese which generally retards the pearlite reaction rather than the pro-eutectoid ferrite reaction.

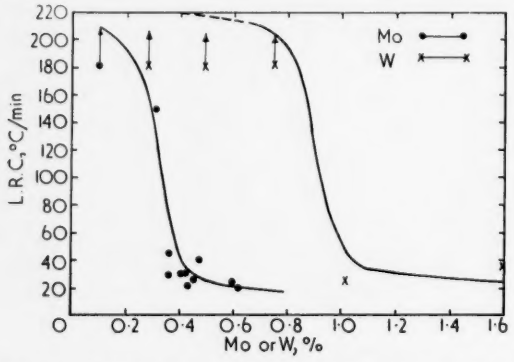


Fig. 7.—Effect of Mo and W on L.R.C. value of low-C, boron steels

(e) Tungsten

In the absence of molybdenum, tungsten causes a pronounced decrease in the L.R.C. but about two to three times as much tungsten is required to produce the same effect as 0·5% Mo. Combinations of 0·25% W with 0·25% Mo can give an L.R.C. value of 25° C/min. Figure 7 shows the very similar effect of tungsten and molybdenum, which might be expected from their general alloying behaviour.

(f) Other Elements

Chromium, nickel, and vanadium have little effect on the L.R.C. value in the absence of molybdenum, although each of them will show some effect when 0·25% Mo is present.

Effect of Analysis on Ac_3 Temperature

It has been reported by several workers that boron does not affect heat treatment temperatures, although it seems to be fairly common practice to treat boron steels at slightly higher temperatures than the corresponding boron-free steel. Ac_3 temperatures for boron steels have been determined dilatometrically at a rate of heating of 5° C/min and results representing about 100 steels of various carbon contents, but containing 0·5% Mo, are shown in Fig. 8. It will be seen that boron not only raises the Ac_3 tempera-

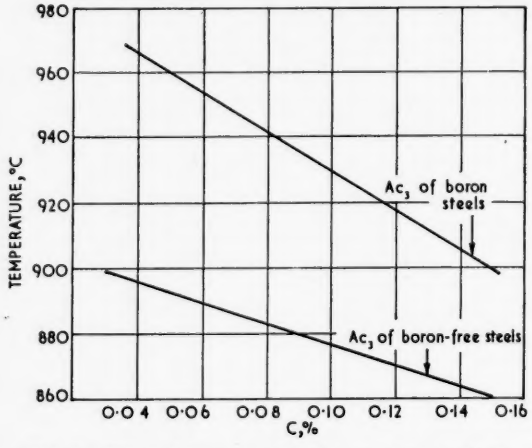
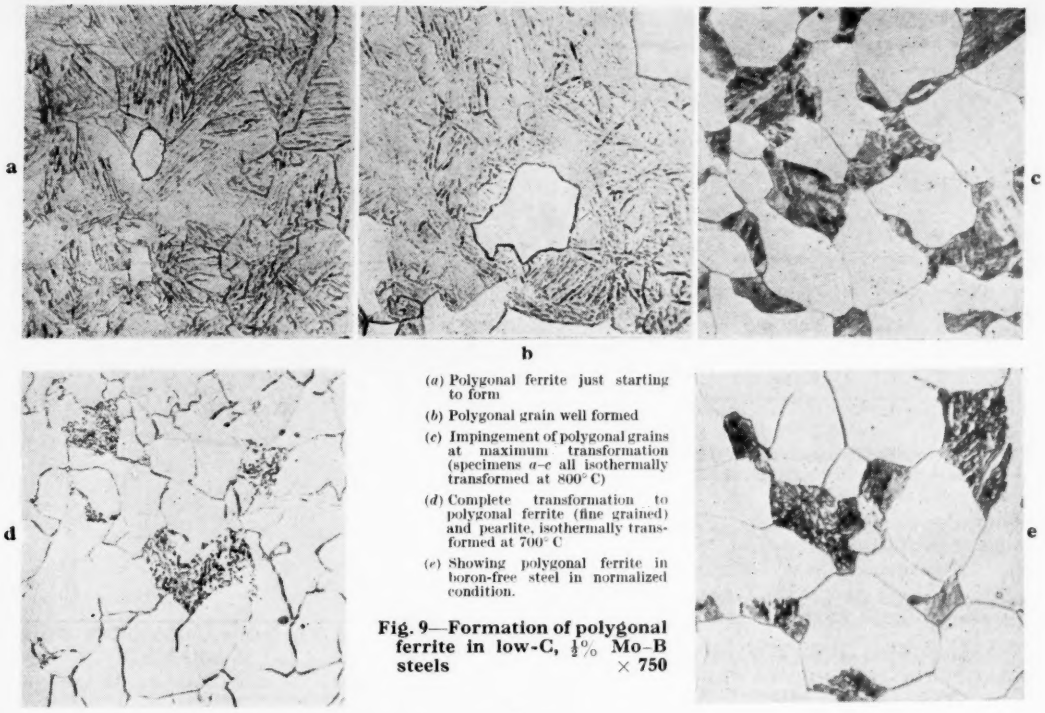


Fig. 8—Effect of boron and carbon on Ac_3 of steels



(*a*) Polygonal ferrite just starting to form

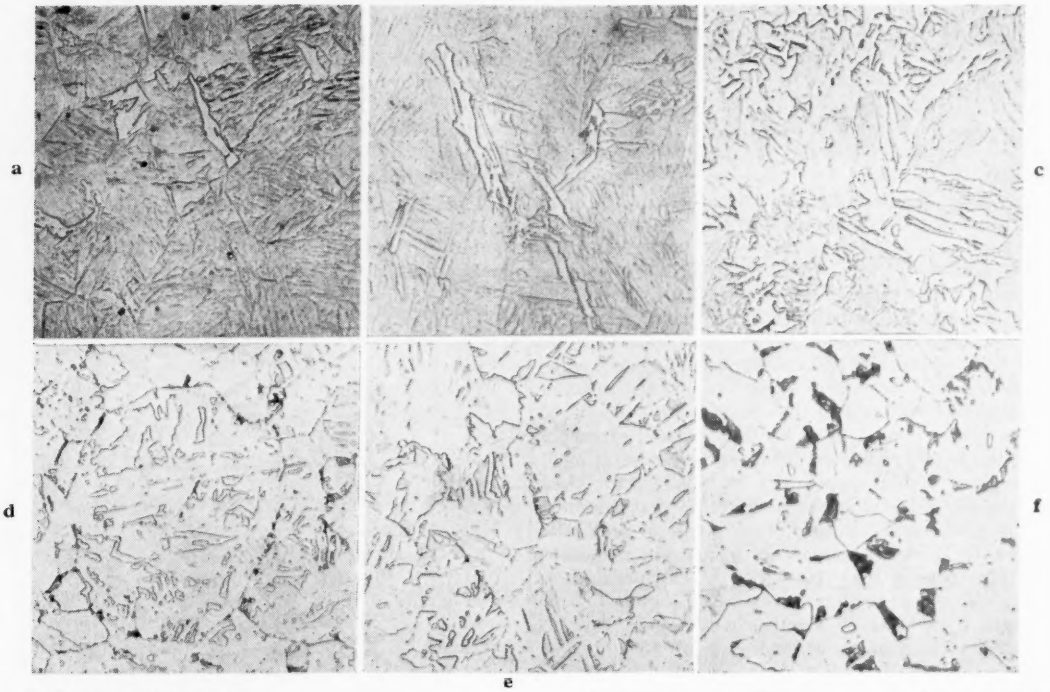
(*b*) Polygonal grain well formed

(*c*) Impingement of polygonal grains at maximum transformation (specimens *a*–*c* all isothermally transformed at 800° C)

(*d*) Complete transformation to polygonal ferrite (fine grained) and pearlite, isothermally transformed at 700° C

(*e*) Showing polygonal ferrite in boron-free steel in normalized condition.

Fig. 9—Formation of polygonal ferrite in low-C, ½% Mo–B steels × 750



(*a*) First plates of bainitic ferrite
(*b*) Growth and formation of new plates of bainitic ferrite
(*c*) Bainitic ferrite now beginning to subdivide remaining austenite into small angular areas

(*d*) Large amount of bainitic ferrite has formed, showing relationship to prior austenite grain outlined by Fe_2B
(*e*) Similar effect to (*d*) but with smaller austenite grain size
(*f*) Bainitic ferrite in normalized steel containing boron (note similarity with (*e*) and compare with Fig. 9*e*)

Fig. 10—Formation of bainitic ferrite in low-C, ½% Mo–B steel, isothermally transformed at 550° C ×750

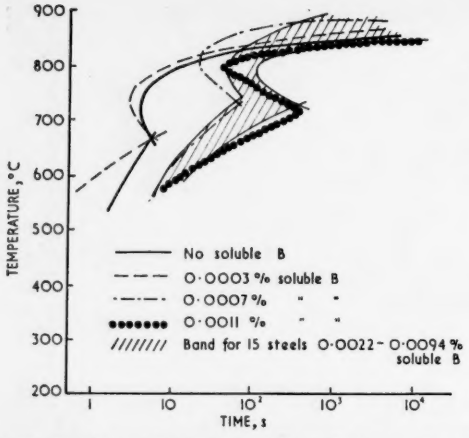


Fig. 11—Isothermal transformation diagrams for ½% Mo steels containing various amounts of boron

ture but also increases the slope of the line. Boron does not seem to have any effect on the Ac_1 temperature. Molybdenum over the range 0·31–0·62% raises the Ac_3 by about 12° C per 0·1% Mo, and silicon also raises the Ac_3 by about 6° C per 0·1% Si.

ISOTHERMAL TRANSFORMATION CHARACTERISTICS OF LOW-CARBON, LOW-ALLOY STEELS CONTAINING BORON

General Transformation Characteristics

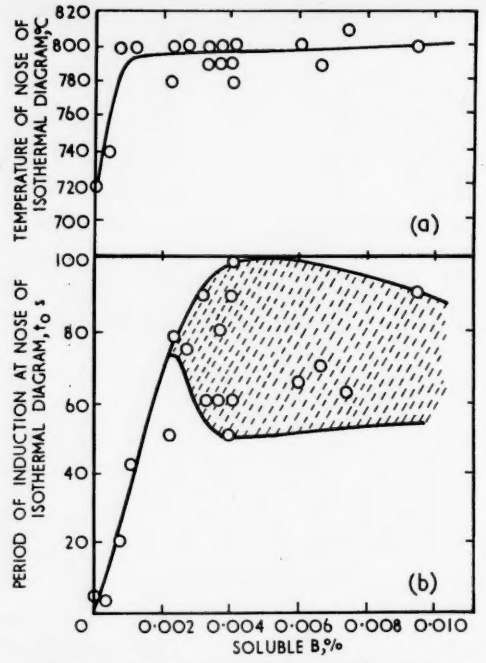
In low-carbon, low-alloy steels, ferrite can form isothermally in two separate temperature ranges. The high-temperature reaction occurs both above and below Ac_1 and produces relatively soft polygonal ferrite. The reaction is one of nucleation and growth, and the various stages of the reaction are shown in Fig. 9. The final polygonal ferrite grains are of the order of 10^{-3} in. dia.

The lower temperature reactions occur in the temperature range from about 700° C down to the M_s and produce a harder, more acicular type of ferrite. These plates of ferrite form within the austenite grains as shown in Fig. 10, and they are at least an order of magnitude smaller than the polygonal ferrite. Electron microscope examination has also shown that each ferrite plate contains sub-grains of even smaller dimensions.[10] According to Ko[11] this transformation takes place by shear and the continued growth of the ferrite plates is controlled by the removal of carbon from the ferrite either by diffusion or precipitation.

In plain carbon steels, the '*C*' curves for these two reactions overlap to such an extent that the shape of the isothermal diagram often gives no indication of the temperature ranges in which the respective reactions take place. However, in the presence of 0·5% Mo, the two reaction '*C*' curves become separated to produce the distinct bay in the 0% transformation line shown in Fig. 2.

Effect of Boron

In the presence of 0·4–0·5% Mo, boron affects the isothermal curve as shown in Fig. 11 and the main features are shown more clearly in Fig. 12. If the hardenability is expressed in terms of the period of



(*a*) Effect of boron on temperature of nose of isothermal diagram
(*b*) Effect of boron on period of induction at nose of isothermal diagram.

Fig. 12—Effect of boron on isothermal transformation characteristics (0·10–0·15% C, 0·4–0·5% Mo, 0·5% Mn)

induction at the nose of the isothermal diagram, then hardenability increases linearly with increasing boron up to 0·002%, above which the period of induction becomes variable, even tending to decrease, although not sufficiently to impair the hardenability. The effect is similar to that shown by the L.R.C. test, and can be explained on the basis that up to 0·002% boron, no supersaturation of the austenite grain boundaries occurs. Above 0·002% B, however, the local solubility at the austenite grain boundaries is

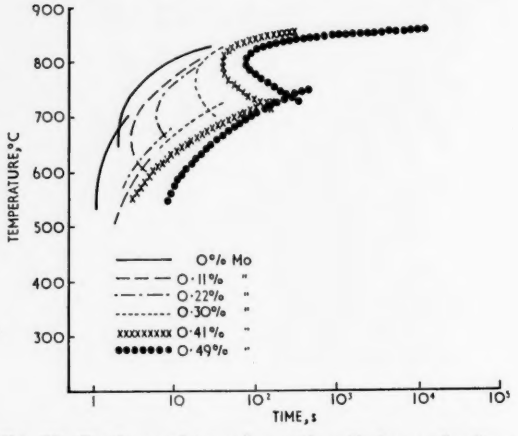
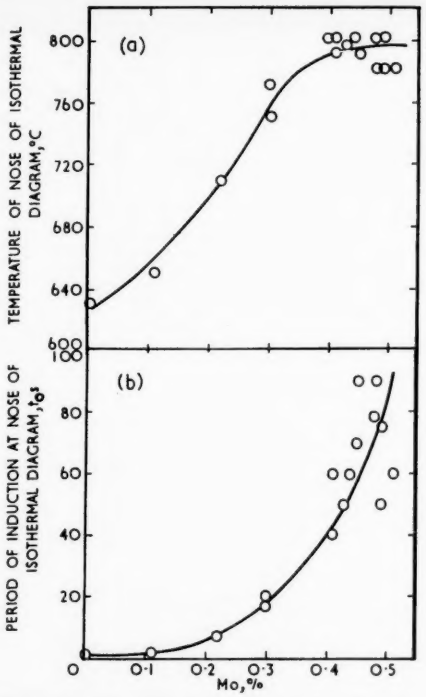


Fig. 13—Isothermal transformation diagrams for low-C, boron steels of various Mo contents (0·10–0·15% C, 0·002–0·0045% B, 0·5% Mn)



(*a*) Effect of Mo on temperature of nose of isothermal diagram
(*b*) Effect of Mo on period of induction at nose of isothermal diagram.

Fig. 14—Effect of Mo on isothermal transformation characteristics (steel as in Fig. 13)

exceeded, with the result that a boron-rich phase precipitates. This phase, which is discussed more fully in a later section of this paper, acts as a nucleus for transformation by providing a phase interface

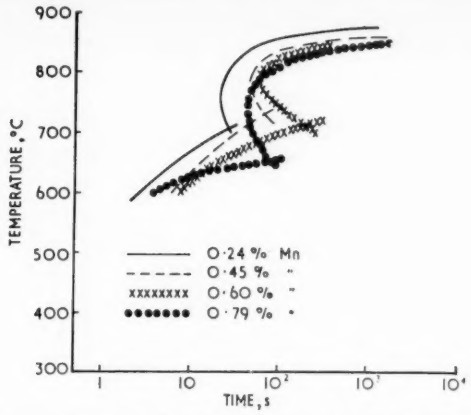


Fig. 16—Effect of Mn on isothermal transformation characteristics of low-C, ½% Mo–B steels

and consequently the period of induction becomes erratic. At about 0·002% soluble boron, the boron constituent becomes visible without a deliberate coarsening of the austenitic grain size.

It has been frequently claimed that boron merely displaces the reaction curves to the right with no appreciable vertical movement. This is not strictly true, since Fig. 12*a* shows that increasing boron contents rapidly raise the nose of the polygonal ferrite '*C*' curve from 720° to 800° C. Above 0·0007% B, however, there is no further effect.

The temperature range of the bainitic ferrite reactions is not affected by increasing boron contents as much as the polygonal ferrite curve and consequently the marked bay is produced in the 0% transformation line. Boron does not appear to have any systematic effect, however, on the completion of transformation.

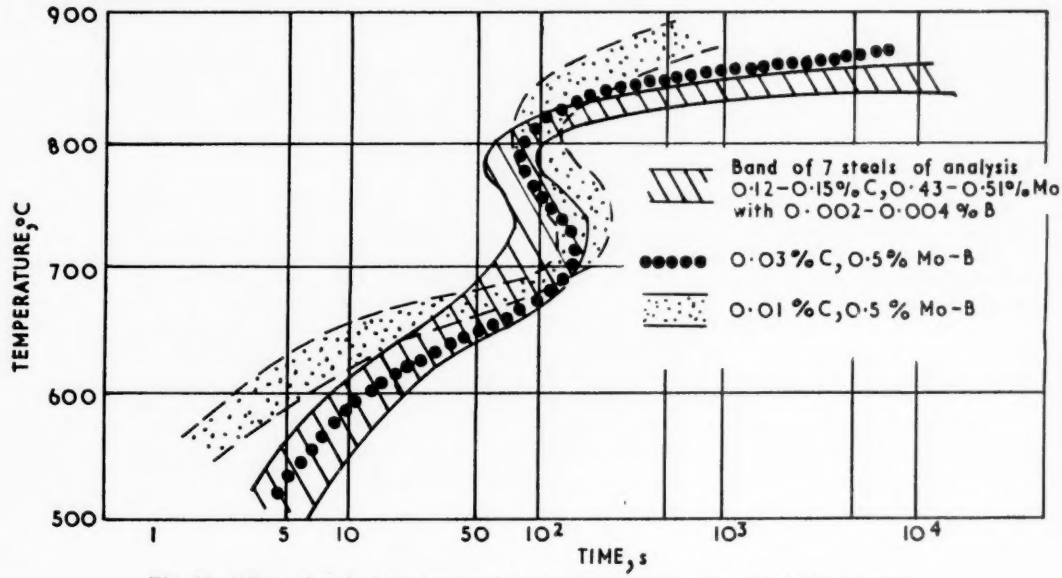


Fig. 15—Effect of carbon content on 0% transformation curve of ½% Mo–B steels

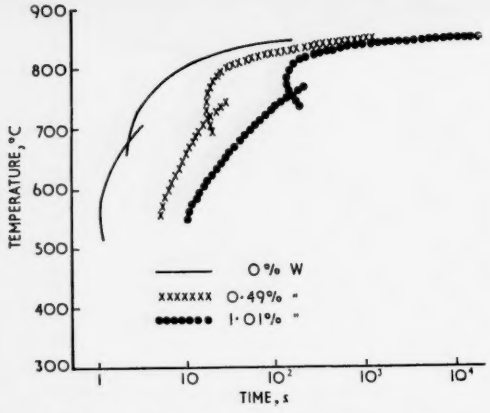


Fig. 17—Effect of W on isothermal transformation characteristics of low-C, boron steels

Effect of Molybdenum

In the presence of sufficient boron to promote optimum hardenability, increasing molybdenum has the effect shown in Fig. 13 and the important details are summarized in Fig. 14. It is clearly shown that increasing molybdenum increases the period of induction at the nose of the isothermal diagram up to at least 0·5% Mo. Molybdenum also raises the temperature of the nose of the polygonal ferrite '*C*' but this effect levels out at about 0·4% Mo. This is in agreement with the general effect of molybdenum in raising the critical points of steels.

Effect of Carbon

The curves shown in Fig. 15 indicate that a variation in carbon content from 0·01% to 0·15% has no significant effect on the general position of the isothermal transformation diagram, providing that adequate (but not too much) boron and adequate molybdenum are present. Indications from work at present in progress are that this is true up to at least 0·2% C and probably much higher providing due regard is paid to the effect of carbon in decreasing the Ac_3 and thereby lowering the temperature of the nose of the polygonal ferrite '*C*' curve.

The small effect of carbon is important from the point of view of the mechanism of the action of boron in promoting hardenability and will be discussed more fully later.

Effect of Manganese

Evidence has been obtained to show that increasing manganese up to 0·4% increases the period of induction at the nose of the isothermal diagram (Fig. 16). Above 0·4%, increasing manganese has little effect. From the isothermal diagrams it would appear that the main effect of manganese is to lower the bainitic ferrite curve, and it continues to do this with manganese contents greater than 0·4%.

Effect of Tungsten

Figure 17 shows that tungsten has a very similar effect to molybdenum, in that the polygonal ferrite '*C*' curve is moved preferentially to the right, but the two '*C*' curves are not separated so clearly as the curves shown in Fig. 13. It will also be noted that two to three times more tungsten is required to produce an effect comparable to molybdenum. The comparison between the effects of molybdenum and tungsten is shown in Fig. 18, which also shows the very small effect of chromium.

Effect of Other Elements

In the absence of molybdenum, chromium in amounts up to 1% has a very small effect although there are some indications from work in hand that a more pronounced effect is obtained with chromium contents of about 3% which then begins to show an effect similar to about $\frac{1}{4}$–$\frac{1}{2}$% Mo.

No isothermal transformation data are available on the effect of vanadium, but from continuous cooling transformation results there is some evidence that in the presence of 0·20% Mo, increasing vanadium up to 0·25% moves the complete isothermal diagram to the right without introducing a bay into the curve, and lowers the transformation temperature below that produced by 0·5% Mo.

'Boron Constituent'

A metallographic test for the presence of boron in steel was devised by Grange[5] who austenitized a steel at 1250° C, then partially transformed it at 650° C. In boron steels a network of fine particles is observed at the prior austenite grain boundaries, and these particles are clearly revealed by the white etching ferrite. This grain boundary phase has been called the 'boron constituent', and this present work has shown that it can be precipitated at any temperature in the range 550–800° C. Typical examples of the boron constituent are shown in Fig. 19.

The idea of the Grange test for boron is mainly to coarsen the grain size, thereby increasing the boron concentration at the grain boundaries to such an extent that the solubility is exceeded and the excess boron is precipitated at the prior austenite grain boundaries. This will obviously depend upon the initial boron content, however, and examples of the boron constituent precipitating at normal grain sizes have been noted. Figure 19 shows that the constituent can form both from ferrite and directly from austenite.

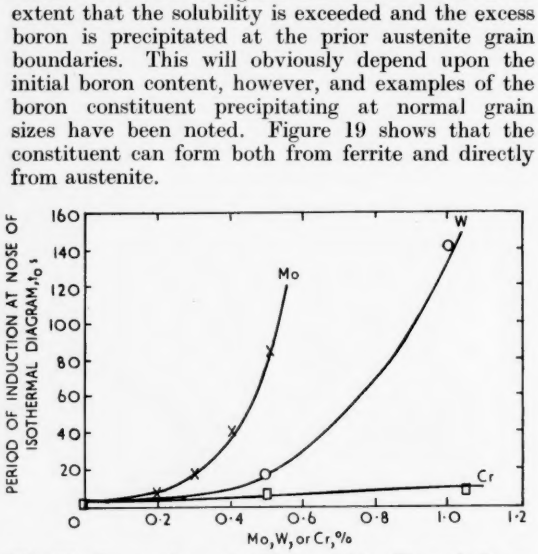
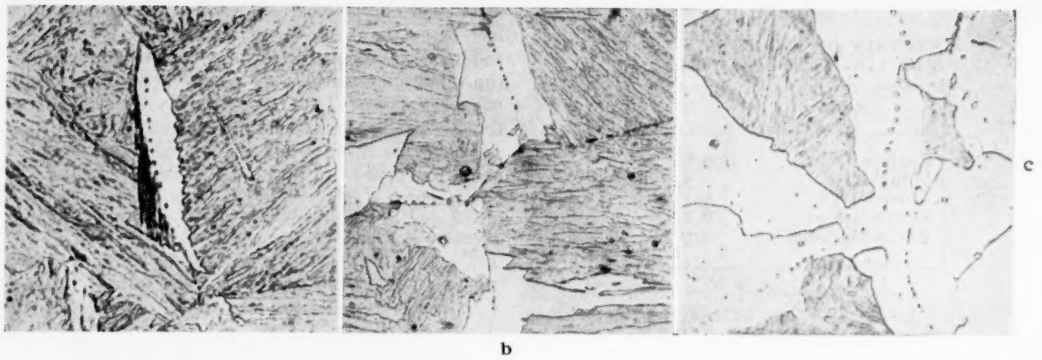
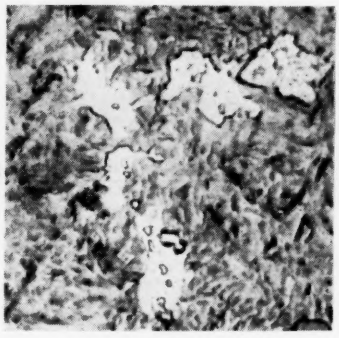


Fig. 18—Comparison of effects of Mo, W, and Cr on isothermal transformation characteristics of low-C, boron steels





(*a*) Fe_2B on grain boundary with ferrite nucleated also at grain boundary

(*b*) Further development of boron constituent

(*c*) Coalescence of boron constituent along prior austenite grain boundary

(*a–c*: Formation of Fe_2B isothermally at 700–750° C in low-C, ½% Mo–B steel × 750)

(*d*) Fe_2B forming mainly in ferrite

(*e*) Fe_2B forming direct from austenite in places

(*d–e*: Formation of Fe_2B isothermally at 600° C in ¼% C, ½% Mo–B steel × 1500)

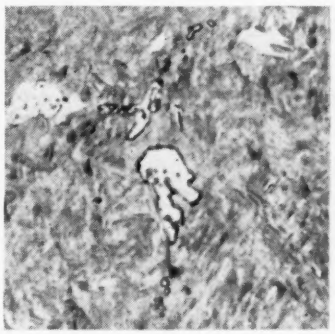


Fig. 19—Some examples of the boron constituent Fe_2B

The nature of the boron constituent has stimulated much interest and early suggestions were made that it was a complex boride–carbide–nitride aggregate. Such suggestions seem very complex, however, since most of the evidence suggests a simple solubility effect. In the steels under examination, nitrogen is fixed by either titanium or aluminium and consequently a nitride is unlikely. More recently, however, electron diffraction work at Cambridge University[10] has identified the boron constituent as Fe_2B.

SOME EFFECTS OF TEMPERING TREATMENTS

A considerable amount of work has been devoted to the study of the effect of time and temperature on the softening characteristics of martensites[12, 13] and on the effects of alloying elements in retarding such softening.[14] Much less information is available on the tempering of intermediate transformation products.

Nutting[15] has shown that a plot of hardness against the logarithm of the tempering time gives a straight-line relationship, and activation energies have been calculated which show that softening during tempering is controlled by grain growth of the ferrite grains. This grain growth is accompanied by coalescence and growth of the carbide particles.

The tempering of bainitic ferrite in low-carbon, molybdenum–boron steels has been investigated, first because it was necessary to know the softening which might be expected during a stress-relieving treatment, but later from a more fundamental standpoint.

The initial work was carried out on a commercial cast of low-carbon, ½% molybdenum–boron steel, details of which are given in Table II.

After normalizing from 970° C, specimens were reheated to temperatures between 500° and 700° C for periods up to 5000 h. A typical series of tempering curves is shown in Fig. 20*a*, which has some interesting features. The initial softening on a logarithmic time scale is linear, which denotes tempering, whilst the original fine-grained bainitic structure persists (Fig. 21*a*). The very rapid softening at a time which depends upon the tempering temperature, corresponds to an equally rapid grain growth of the bainitic structure into polygonal grains (Fig. 21*b*), with the accompanying formation of coalesced particles on the prior austenite grain boundaries. These particles may be either carbide or Fe_2B which might be expected to precipitate at the initial site of the austenite grain boundary. The slight softening below the knee of the curves in Fig. 20*a* may well be due to slight grain growth of the polygonal ferrite.

A better analysis of the results can be made by using the combined time/temperature parameter developed by Holloman.[13] Such an analysis gives the band shown in Fig. 20*b* from which the value of $T(20+\log t)$ can be assessed for the rapid softening reaction. From this value it is possible to determine a safe tempering time at any temperature. Also it was readily apparent that the kinetics of the process could be studied by short-time treatments at the higher temperatures around 600–700° C, rather than by long times at lower temperatures.

Table II

ANALYSIS OF STEEL USED IN INITIAL EXPERIMENTS

Element	%
Carbon	0·10
Manganese	0·52
Silicon	0·174
Molybdenum	0·52
Boron (soluble)	0·0022

The investigation was then extended to determine the tempering characteristics of W.Q. (i.e. martensitic) material as compared with normalized (bainitic) material.

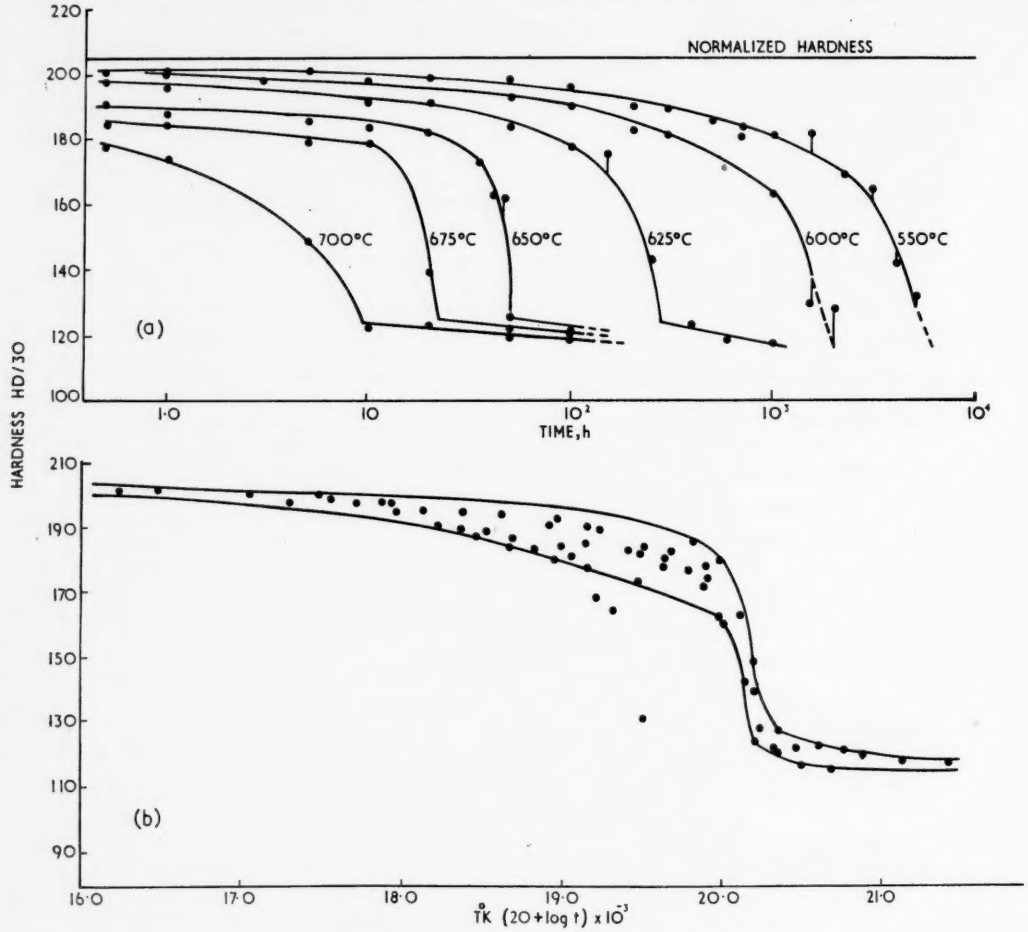
The changes in structure and hardness were similar in character in both the normalized and W.Q. structures, the initial softening being linear on a logarithmic time scale, followed by a rapid decrease in hardness and the final slow softening. Again the rapid softening was associated with grain growth of the ferrite.

The W.Q. specimens always softened more rapidly than the normalized specimens, (compare Fig. 22 with Fig. 20*b*). This was probably because the martensite was less stable than the bainitic ferrite.

Finally, a marked age-hardening effect was sometimes observed in both the normalized and W.Q. specimens, as shown in Fig. 22. Such an ageing effect might be due to the precipitation of carbide or Fe_2B during tempering.

MECHANICAL PROPERTIES OF LOW-CARBON, MOLYBDENUM-BORON STEELS

The mechanical properties of this type of steel have been described in detail by Bardgett and Reeve.[1, 16] The most important point is that, provided the high-temperature ferrite formation is suppressed, then improved properties, particularly tensile strength, are obtained from the bainitic ferrite structure which is formed. The close relationship between the mechanical properties and the trans-



(*a*) Family of isothermal time/hardness tempering curves
(*b*) Holloman plot of data in (*a*)

Fig. 20—Tempering characteristics of low-C, ½% Mo–B steel normalized at 970° C

formation data is shown in Fig. 23. In each case the effect of the alloy element on the mechanical properties compares directly with the effect of that element on the L.R.C. value and the transformation diagram.

Welding Characteristics

Bardgett and Reeve showed that ½% molybdenum–boron steel possessed good weldability with freedom from hard zone cracking. This was unusual for a steel with such high mechanical properties. Cottrell[17] has shown that hard-zone cracking does not occur if the heat-affected zone of the weld transforms above 290° C. Generally speaking, however, steels which give mechanical properties as high as the ½% Mo–B steel (i.e. 30 tons/in² per 0·2% proof stress) transform below 290° C, whereas the ½% Mo–B steel transformed at 430° C under welding conditions.

When these results are considered in relation to the general transformation characteristics of ½% molybdenum–boron steels which have been described in this paper, it is easy to appreciate the fundamental difference between the two types of steel. With a steel having a ferrite–pearlite structure, increased strength can only be obtained by increasing the carbon content or by employing solid solution hardening. Either method will lower the M_s point and hence cause transformation to occur below 290° C so that cracking occurs.

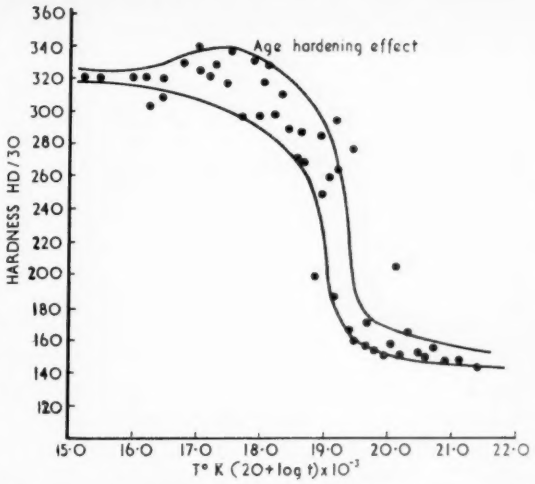
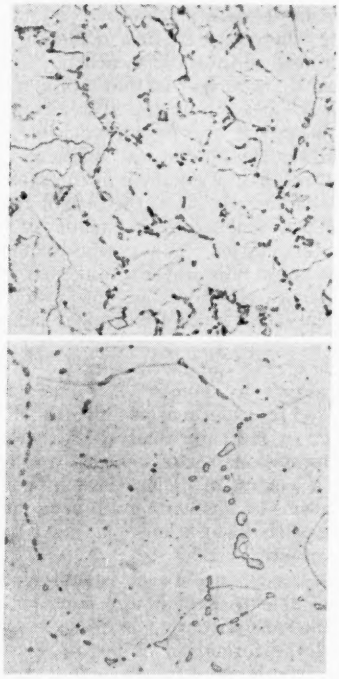


Fig. 22—Tempering characteristics of low-C, ½% Mo–B steel after water-quenching

The high strength of the ½% molybdenum–boron steel, however, is due to the bainitic ferrite structure, which forms over a wide range of cooling rates. Consequently, the transformation temperature is high even under the most severe welding conditions, and cracking is eliminated. The trend in high-strength weldable steels will be towards the development of steels in which the plate, weld metal, and heat-affected zones all have bainitic ferrite structures which have high transformation temperatures and consequently show no hard zone or weld cracking.[18]



(*a*) 20 min at 700° C, showing original bainitic ferrite structure with some coalescence of carbide
(*b*) 100 h at 700° C, showing polygonal grains of ferrite and coalescence of carbide (or Fe_2B) at prior austenite grain boundaries which often do not coincide with ferrite grain boundaries

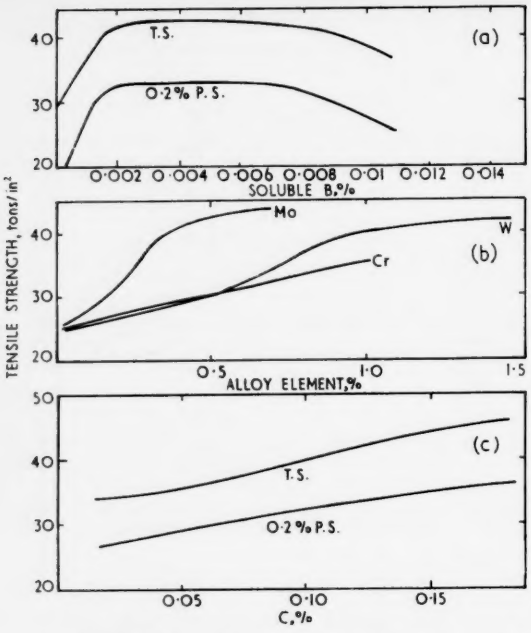
Fig. 21—Structural changes during tempering low-C, ½% Mo steel

DISCUSSION

Some discussion of the results has been given in previous sections of this paper, but two important features which require more adequate explanation are the strength of the bainitic ferrite and the mechanism by which boron increases the hardenability.

The most interesting feature of the strength of the structure which forms when polygonal ferrite is suppressed is the relatively small effect of carbon. The increase in strength that is produced by varying the carbon content from 0·01% to 0·20% is probably due to the increase in the amount of bainitic carbide present in the structure. However, it only amounts to an increase in the 0·2% proof stress of 5–6 tons/in² over the whole carbon range, whereas at a constant carbon content the change from polygonal to bainitic ferrite doubles the 0·2% proof stress, i.e. from 15 to 30 tons/in². It would appear, therefore, that the bainitic carbide contributes little to the strength of the structure.

It does not seem likely that supersaturation of the ferrite with carbon has any effect. There is no evidence that any precipitation of carbide takes place from ferrite during isothermal transformation or tempering; indeed, the mechanism of bainite formation proposed by Ko[11] would indicate that at such a high temperature of formation, carbon could easily



(*a*) Effect of soluble boron content on steels containing 0·13–0·15% C and 0·40–0·60% Mo
(*b*) Effect of alloy content on steels containing 0·10–0·15% C and 0·002–0·007% B
(*c*) Effect of carbon content on steels containing 0·4–0·5% Mo and 0·002–0·007% B

Fig. 23—Effect of composition on mechanical properties

diffuse away from the growing ferrite into the remaining austenite.

Having discounted any effect due to carbon, the possible reasons for the high strength of ferrite are:

(i) Low transformation temperature
(ii) Internal stress (strain induced hardness)
(iii) Fine grain size.

The first two effects are very similar but there are strong reasons for believing that they cannot be important. Since the transformation temperature (although low for ferrite formation) is still relatively high, any hardening due to volume changes or coherency stresses between the matrix and ferrite must be quickly relieved. Also, since the structure is very resistant to tempering it must be concluded that stress- or strain-induced hardness is not the major factor.

The fine grain size appears to be the most important factor. The individual ferrite grains are themselves very small, but electron-microscope studies at Cambridge University have shown that each bainitic ferrite grain contains a pronounced sub-grain structure, and the present work has shown that there is a very marked relationship between the hardness and the size of these sub-grains, with the hardness increasing with decreasing sub-grain size.[10]

It seems likely, therefore, that the high strength of the bainitic ferrite depends upon the small sub-grain size, but this still does not explain how the sub-grain structure is produced. It may be produced by the coherency strains developed during the formation of bainitic ferrite (assuming Ko's hypothesis). This fine grain size would then be very resistant to tempering since any softening would require grain growth.

Turning now to the mechanism of the effect of boron on hardenability, the generally accepted ideas are that boron segregates to the austenite grain boundaries and in so doing decreases the rates of nucleation for transformation at the grain boundaries. It has already been suggested that since boron has little effect in high-carbon steels, and the effect of carbon on the transformation diagram of the present steels is virtually nil, it is only the grain-boundary nucleated ferrite reaction which is retarded by boron. The formation of ferrite at an austenite grain boundary requires a nucleus of a ferrite lattice to be produced of such a size that the thermodynamic conditions for stability and growth are fulfilled. Little supersaturation with carbon exists at the temperatures at which polygonal ferrite forms, and therefore the controlling factor for the nucleation and growth of ferrite is the rate of diffusion of carbon into the remaining austenite. The formation of a ferrite nucleus does not necessarily involve the diffusion of boron, since boron is more soluble in ferrite than austenite.[19]

Carbon segregates to austenite grain boundaries in a similar manner to boron[3] but, normally, ferrite has no difficulty in forming a nucleus and growing because of the rapid diffusion of carbon. Often the ferrite grows preferentially along the boundary because carbon can most easily diffuse away along the grain boundary. There is little firm evidence that boron reduces the rate of diffusion of carbon. The mechanism envisaged, however, is that boron reduces the diffusivity of carbon near the austenite grain boundaries by occupying the most favourable interstices, thus retarding the interstitial diffusion of carbon. This would result in a decrease in the rate of nucleation but would hardly affect the rate of growth, since boron is mainly confined to the vicinity of the austenite grain boundaries. Other workers[20] have suggested that boron relieves the strain energy due to the austenite–ferrite reaction, but this would not seem to apply since polygonal ferrite does not form by a process of lattice shear. However, such a mechanism could explain the slight retardation of the upper bainite reaction by boron. A further hypothesis[2] has suggested that boron atoms fill sites of high strain energy thereby reducing their effectiveness as nuclei for transformations, whilst other workers[21] have suggested a locking of dislocations by concentration of boron atoms at austenite grain boundaries, hereby preventing such dislocations playing a part in the formation of ferrite nuclei.

No matter what mechanism is envisaged, however, it must be able to explain the decreasing effect of boron as the carbon content increases. Carbon itself will retard the formation of ferrite, because as the average carbon content increases it becomes increasingly difficult for a nucleus of critical size and free from carbon to form. Boron acts in a very similar manner and hence its effect would be expected to decrease as the carbon content increases. Spretnak and Speiser[20] also conclude that boron acts on the rate of nucleation of ferrite in the same way as carbon, and they indicate a similar effect with nitrogen.

The effect of boron content and grain size can be summarized briefly. At any particular grain size, the ferrite nucleation rate will decrease with increasing boron content until the local solubility for boron is exceeded at the grain boundaries. At higher boron contents Fe_2B precipitates and the nucleation rate becomes erratic. Increasing austenitizing temperature has several effects. In the first place a coarser grain size increases the boron content at grain boundaries, but the higher temperature can also allow the boron to diffuse away from grain boundary regions by increasing the thermal activation. It is also possible that if nitrogen has been fixed with aluminium, high austenitizing temperatures will cause dissociation of aluminium nitride, leaving the nitrogen free to combine with the boron. Finally, very high boron contents ($> 0{\cdot}007\%$) may have a bad effect on hot workability because of the continuous films of Fe_2B eutectics at austenite grain boundaries.

SUMMARY AND CONCLUSIONS

(1) The transformation characteristics of low-carbon low-alloy steels containing boron have been examined isothermally and on continuous cooling. Boron has been shown to exert a maximum effect at about the solubility limit for Fe_2B at the austenite grain boundaries and a hypothesis for the mechanism by which boron promotes hardenability has been developed.

(2) Of the alloying elements, molybdenum up to $0{\cdot}5\%$ has been shown to have the major effect when present with boron, which is to retard polygonal ferrite formation and to a lesser extent the upper bainitic ferrite formation. This produces a marked bay in the isothermal transformation diagram and allows bainitic ferrite to form over a wide range of cooling rates. Only tungsten shows a similar effect, but more tungsten is required and a less distinct bay is formed on the isothermal transformation diagram. Carbon has very little effect on the transformation diagram over the range $0{\cdot}01$–$0{\cdot}20\%$ C.

(3) The mechanical properties of the steels are directly related to the type of transformation product and the improved properties are obtained whenever the transformation to bainitic ferrite occurs.

(4) The steels are very resistant to tempering but final softening occurs fairly rapidly and this is associated with grain growth of the bainitic ferrite.

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A High-temperature X-ray Study on High-speed Steel

By H. J. Goldschmidt, D.Sc., F.Inst.P.

Part I—THE LATTICE EXPANSION OF MATRIX AND CARBIDE

SYNOPSIS

The lattice expansions of the phases coexisting in an 18/4/1 type high-speed steel, namely the ferrite or austenite matrix and η-carbide, have been determined by the high-temperature X-ray method. The resulting coefficients are interpreted in terms of 'physical' and 'chemical' (or solute) expansion, the latter being pronounced in austenite. A point of interest is that the coefficients of η-carbide are close to those of ferrite in iron and steel and higher than those of other hard carbides. The lattice coefficients agree closely with those known from bulk dilatometry; the latter are governed by the matrix expansion but, for the two-phase aggregate of high-speed steel, can be affected to a notable extent by the carbide. The relative carbide and ferrite expansions are briefly considered theoretically in the light of Grüneisen's law. Part II will deal with isothermal transformations during tempering and with non-equilibrium aspects. 1305

ONE OF THE ATTRACTIONS of the high-temperature X-ray diffraction method is that it permits one to observe not only the expansion of a single-phase substance, but also the separate expansions of each constituent in a two- or multi-phase assembly, where normal dilatometry gives only the bulk effect.

It is possible, therefore, in an alloy to compare the coefficients of thermal expansion of two coexisting phases at varying temperatures and to study any compositional interaction that may occur between the constituents, i.e. what might be termed the 'chemical,' as distinct from the 'physical,' expansion. In addition any structure transformations may be seen, such as the α/γ change in a steel.

High-speed steels, apart from the intrinsic interest in their high-temperature structures, lend themselves well to illustrate the principles of lattice expansion measurements in multi-phase materials, because the α- or γ-matrix phases and the main carbide constituent, η, possess well-defined diffraction patterns persisting at all solid temperatures. The object of this investigation was to study the expansion of the ferrite, austenite, and η-carbide phases in a normal 18/4/1 (W–Cr–V) type high-speed steel in the as-annealed condition for temperatures up to about 1100° C. A further development would be the isothermal study of the (non-equilibrium) kinetics of the martensite transformation and one aspect of this, that of tempering by means of time-sequence photographs, will be dealt with in a second paper.

The nature of high-speed steels and their carbides is well known from previous work and reference to some of the extensive literature may suffice.[1–12] This includes X-ray room-temperature work on the steels and on carbide extracts[6, 9–12] but, to the best of the author's knowledge, a high-temperature X-ray study has not before been made.

The principal and often exclusive carbide is the η-carbide, of composition M_6C, where M = (Fe, W; minor Cr, V), i.e. Westgren's Fe_4W_2C/Fe_3W_3C. Further trace phases are liable to occur under special conditions, namely, chromium-rich carbide of the $M_{23}C_6$ type, VC and Fe_3W_2; the latter compound occurs when the steel is decarburized. Additionally W_2C, Fe_3C and other non-equilibrium carbides have been detected by some investigators.

EXPERIMENTAL

(*a*) Material

The material used was an 18/4/1 type high-speed steel of the following composition (weight-%): 18·3 W, 4·15 Cr, 1·10 V, 0·73 C, approximately 0·4 Mo, 0·25 Mn, 0·15 Si, balance Fe. The steel, received in bar form in the fully annealed condition, was machined to a thin wire approximately 2 cm long and ½ mm diameter. After being annealed this wire was used as specimen in the high-temperature powder-camera. The state of heat treatment of the sample previous to the X-ray exposure is, of course, an important factor and the aim was to obtain equilibrium as nearly as practicable, both initially and at temperature except in certain cases to be discussed, and to relieve stresses introduced during preparation.

The wire specimens were annealed at 700° C *in vacuo*; the time-periods differed from case to case, as will be mentioned later. It was confirmed that the initial constitution of all specimens was pure, strain-free ferrite with η-carbide.

(*b*) X-ray Technique

The work was carried out using the B.S.A. high-temperature X-ray camera (second model) which has been described previously.[13, 14] High vacuum was maintained throughout, the pressure being between 0·05 and 0·2 μ of mercury (measured at the vacuum exit end of the chamber). The temperature was determined by means of a Pt/Pt–13% Rh thermocouple closely adjacent to the sample. As discussed elsewhere[15] special attention was paid to temperature calibration which was based on known lattice parameters and dilatometric data for pure silver, tungsten, and platinum. Pure iron was examined in parallel, under conditions identical with those for the high-speed steel sample. (This formed an investigation to be reported separately.[16] Lattice expansion and A_3-point results agreed closely with the best available data, and with recent work from Hume-Rothery's laboratory.[17]) The data on α and γ in pure iron were

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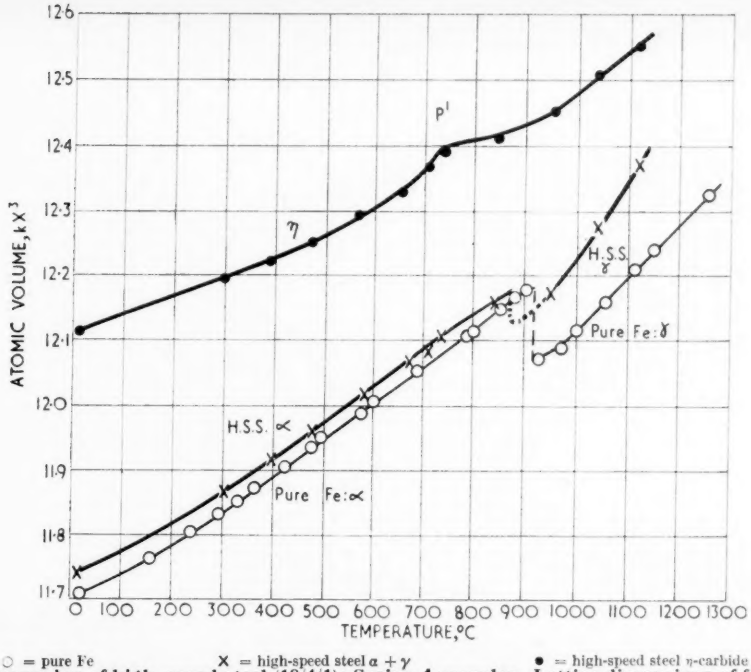


○ = pure Fe X = high-speed steel $\alpha + \gamma$ ● = high-speed steel η-carbide

Fig. 1—Lattice expansion of high-speed steel (18/4/1); Series *A* samples. Lattice dimensions of ferrite, austenite, and η-carbide as a function of temperature. Measurements for pure Fe are included for comparison

Table I

CONSTITUTION, LATTICE CONSTANTS AND ATOMIC VOLUMES OF α-, γ-, AND η-PHASES IN HIGH-SPEED STEEL, AS A FUNCTION OF TEMPERATURE

Temperature, ° C	Constitution	Lattice constant, kX	
		α and γ	η
SERIES A			
10	$\alpha + \eta$	2·8637	11·064
300	,,	2·8735	11·093
392	,,	2·8775	11·103
473	,,	2·8810	11·113
578	,,	2·8857	11·125
663	,,	2·8895	11·138
707	,,	2·8908	11·145
735	,,	2·8928	11·154
845	$\alpha + \eta$ + small amount of γ	2·8968	11·159
952	$\gamma + \eta$	3·6512	11·172
1039	,,	3·6615	11·189
1117	,,	3·6708	11·200
SERIES B			
10 (7 days at 700° C)	$\alpha + \eta$	2·8633	11·064
10 (28 days at 700° C°)	,,	2·8632	11·066
207	,,	2·8709	11·087
289	,,	2·8727	11·097
456	,,	2·8790	11·116
621	,,	2·8857	11·135
737	,,	2·8915	11·150
849	$\alpha + \eta$ + small amount of γ	2·8971	11·163
893	$\gamma + \eta$	3·6486	11·169
951	,,	3·6529	11·173
1052	,,	3·6648	11·190

used for comparison with the corresponding data on high-speed steel.

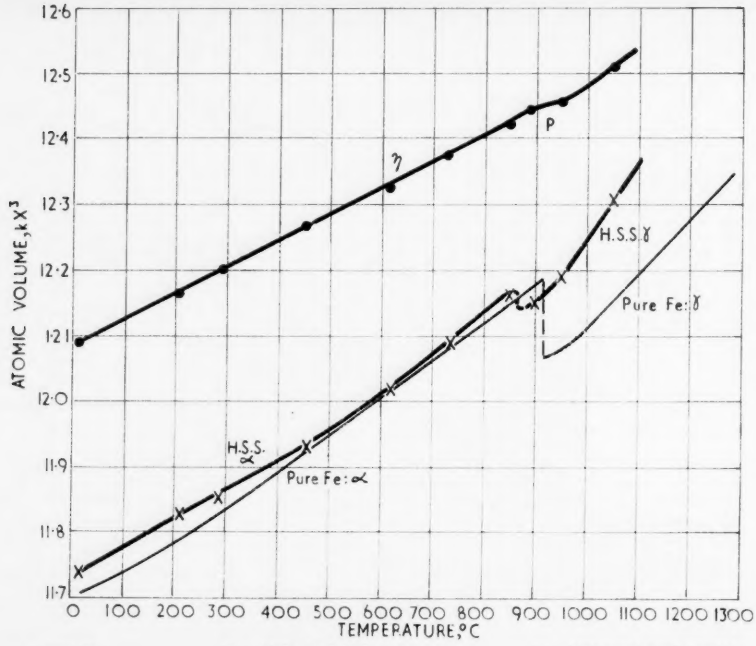
The accuracy of temperature determination in the present high-speed steel work was approximately $\pm 2°$ C. Cobalt $K\alpha$ radiation was used throughout, the wavelengths used being: Co $K\alpha_1 = 1{\cdot}78529$kX, Co $K\alpha_2 = 1{\cdot}78919$ kX. All lattice parameters will be expressed in (Siegbahn) kX units. They were measured by the usual extrapolation method, principal use being made of the high-angle reflections (310) for ferrite, (400) for austenite, and (12,3,0) and (12,2,0) (i.e. $\Sigma h^2 = 153$ and 148) for η-carbide. The camera diameter was 19 cm.

RESULTS

The constitutions and lattice dimensions obtained at each temperature are shown in Table I. The results are expressed in terms of lattice spacing a, and in Figs. 1 and 2 of the mean atomic volumes A; the latter facilitate direct dimensional comparison between the three different structures. The relationships between a and A are for α: $A = a^3/2$; for γ: $A = a^3/4$; and for η-carbide: $A = a^3/112$ (the 112 atoms per unit cell in η comprising 96 metal and 16 carbon atoms).

In Figs. 1 and 2, the results are shown alongside those obtained on pure iron, in the parallel investigation.

There are two experimental series A and B with a certain difference in the pre-heat-treatment of the samples. In Series A each sample was held at 700° C for 2 hours before the first X-ray exposure, in Series B each sample was held at 700° C for a prolonged time, namely for 7 days and as additional check for 28 days. This latter extra time had no effect, the 7 days



— = pure Fe × = high-speed steel $\alpha - \gamma$ ● = high-speed steel η-carbide

Fig. 2—Lattice expansion of high-speed steel (18/4/1); Series *B* samples. Lattice dimensions of ferrite, austenite, and η-carbide as a function of temperature. Measurements for pure Fe are included for comparison.

having in practice already produced equilibrium, but the difference between 2 hours and 7 days mattered significantly.

The coefficients of expansion for the α-, γ-, and η-phases are given in Figs. 3 (mean) and 4 (true) coefficients. These apply to Series *B* throughout, as it is fundamentally correct to base the coefficients on data obtained on samples as nearly as possible in equilibrium. (It is a debatable point however whether the expansion coefficients should be based on the 'ideal' equilibrium dimensions of a phase, as attempted here, or on a condition corresponding to Series *A*, which would be more representative of metallurgical practice, Such a basis for evaluation would, however, seem arbitrary and not justified).

The value shown in Fig. 3 is $\beta_m = [(a_T - a_0)/a_0]/(1/\Delta T)$ namely, the mean coefficient as from room temperature; that in Fig. 4 is $\beta_t = (\delta a/\delta T)(1/a_T)$ the coefficient at the temperature given. For temperatures in the γ-region the " mean " coefficient as from room temperature is necessarily of limited significance, and has here been formally derived by way of the atomic volumes for the body-centred and face-centred cubic unit cells. Data within the *austenite region only* are given in Table II.

It may be seen that the final coefficients at the high-temperature end agree well in Series *A* and *B*. The true coefficients β_t at 1050° C are 33·3 (γ), 16·6 (η) in Series *B*, and 32·8 (γ), 16·6 (η) in Series *A* ($\times 10^{-6}$)/° C.

DISCUSSION

(1) Ferrite and Austenite Expansion

The α- and γ-matrix phases in high-speed steel have wider lattice dimensions than in pure iron, due to solid solution of alloying elements. The expansion coefficients for ferrite tend to be lower in steel than in iron, consistent with accepted ideas (e.g. Fe: $14{\cdot}3 \times 10^{-6}$; high-speed steel: $12{\cdot}9 \times 10^{-6}$ for 10–600° C range).

Austenite in high-speed steel expands nearly linearly, as in pure iron, but with a considerably greater coefficient (Fe: $23{\cdot}3 \times 10^{-6}$, against high-speed steel: $32{\cdot}3 \times 10^{-6}$ for 950–1050° C range). Here an increasing solid solution of carbon and metal atoms clearly superimposes upon the physical expansion as is reflected both by absolute spacings and by coefficients. The source of this progressive solution is the previously free carbide, namely η, plus any subsidiary $M_{23}C_6$ and VC that may have been present.

Comparison of Series *A* and *B* indicates that ferrite has a slightly lower mean rate of expansion in the sample nearer equilibrium (*B*). In Series *A*, the expansion curve for α is virtually linear and runs almost parallel to that for α in pure iron; in Series *B*, on the other hand, there is a slight curvature. This may have been masked in Series *A* by solution of metastable carbides. The austenite results for Series *A* and *B* are in agreement, except for a small difference in curvature at lower temperatures, indicating that in the Series *B* specimens slightly more

Table II

COEFFICIENTS OF EXPANSION INSIDE γ-FIELD

Temperature range, ° C	$\beta_m \times 10^6$/° C Austenite	η-carbide
900 to 1050	28·5	$12{\cdot}0_0$
950 to 1050 (Series B)	32·3	$15{\cdot}2_1$
950 to 1050 (Series *A*)	31·2	$15{\cdot}4_0$

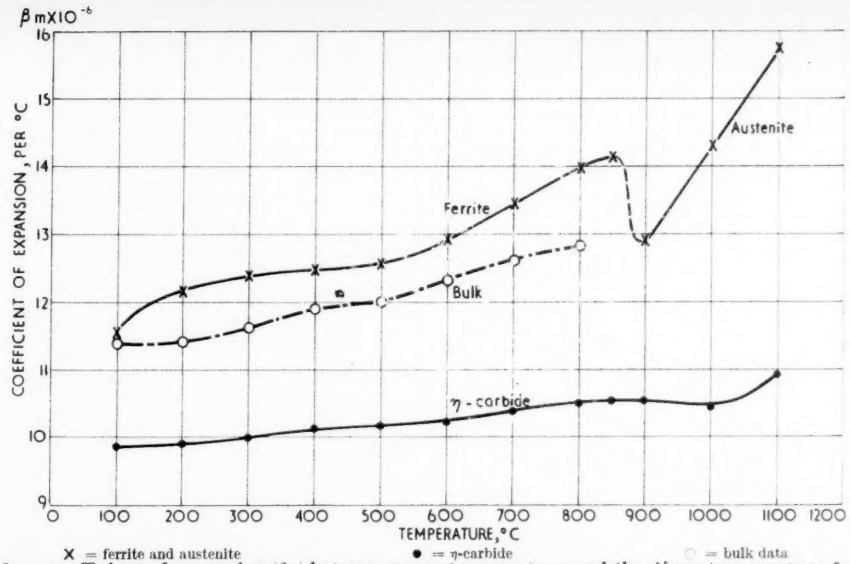


Fig. 3—Mean coefficient of expansion (β_m) between room temperature and the given temperature for ferrite, austenite and carbide in high-speed steel, annealed. Series *B* samples

carbide had been taken into solid solution during preliminary treatment.

Transformation Point—The transformation point $(\alpha + \gamma) \rightarrow \gamma$, i.e. the boundary separating the $(\alpha + \gamma + \eta)$ and $(\gamma + \eta)$-phase fields, was here observed to lie between 850° and 890° C. It was not part of the present work to locate it more accurately. According to Murakami and Hatta's[4] phase diagram (see Part II of this paper) for high-speed steels, recently supported by Kuo,[11] the same change point is located, for the present 0·73% C content, at 880° C in satisfactory agreement.

(2) η-Carbide Expansion

The η-carbide expands virtually linearly throughout, except for a small discontinuity P between 900° and 950° C (Fig. 2). The coefficient of expansion β, accordingly, is almost constant, β_t rising very slowly from 10 to 11 $\times$ 10^{-6} but more rapidly above point P. In Series A the expansion curve shows a slight initial convexity and a marked "hump" P' at 750–800° C, above which the change is practically linear (*see* Fig. 1). It is interesting to note the merging of the curves for Series A and B above about 950° C, i.e. a temperature at which the lack of equilibrium, which was liable to persist at lower temperatures, would be eliminated. It may be concluded that the discontinuity P in Series B is a residue of P' in Series A and that both are caused by essentially the same type of lattice occurrence. That this hump is a metastable effect is demonstrated by the fact that the longer pre-anneal has nearly eliminated it: the probable cause is interaction between η and the coexisting α- and γ-matrix phases and possibly with the minor carbides. It is noted that both the anomalies P and P' in the η-expansion occur in the region of the α/γ change; a connection is not really surprising in view of the greatly different solubility of elements in the α- and γ-phases and it is indeed more noteworthy that the carbide expansion is affected so little by matrix reactions.

The question arises whether, as some η-carbide dissolves in austenite during the $\alpha \rightarrow \gamma$ change, this involves (*a*) a composition change of the carbide, or (*b*) a decrease in total quantity of carbide only, leaving the remaining free carbide substantially unaltered in composition. From the above it would follow that (*b*) is essentially true. Some composition change within the η-carbide mass is, however, not excluded and the rather more rapid expansion of η at the higher temperatures may be indicative of such change.

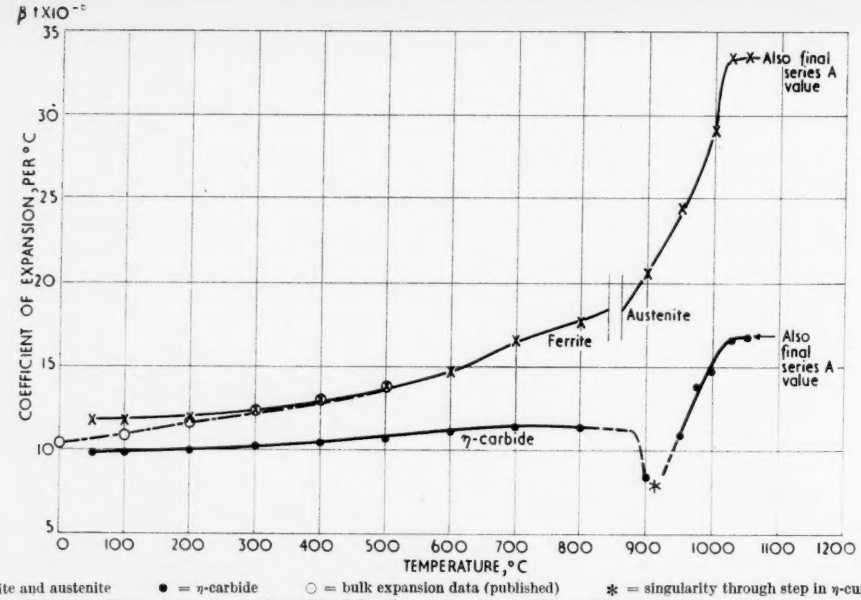
To test this point, an electrolytic carbide extract from the as-hardened steel was prepared for com-

Table III

DETAILS OF CARBIDE EXTRACTS FROM 18/4/1 STEEL

	Composition, wt-%						Constitution	
	Fe	W	Cr	V	Mo	C		
Annealed	24·2 2·1	60·6 1·6	7·4 0·7	4·0 0·37	1·4 0·06	2·5 1	η + small amount $M_{23}C_6$	No change in lattice parameter observed
Quenched (oil, 1300° C)	18·7 1·85	72·2 2·15	2·5 0·25	3·0 0·32	1·4 0·08	2·2 1	η only	

(Underlined figures: atomic compositions, per carbon atom)



× = ferrite and austenite ● = η-carbide ○ = bulk expansion data (published) ✱ = singularity through step in η-curve, Fig. 2

Fig. 4—True coefficient of expansion (β_t) of ferrite, austenite, and carbide in high-speed steel, annealed. Series *B* samples

parison with one from the annealed steel; these were examined chemically and by X-rays. The results are given in Table III.

Chemical analysis therefore reveals a significant composition difference, particularly in chromium content; however this can largely be accounted for by the small amount of extra carbide $M_{23}C_6$ found by X-ray examination of the annealed sample. The η-carbide in the two samples had the same lattice parameters. Thus a small composition change within the η-carbide may superimpose at high temperatures showing slight preferential solution, but the predominant effect is absorption of entire carbide grains. The amount of solution taking place can be estimated from the X-ray line intensities of η- and matrix-phases as shown by the high-temperature photographs. The approximate proportions are:

$$\text{austenite: } \eta \simeq 4{\cdot}5 : 1,$$
$$\text{ferrite: } \eta \simeq 2 : 1.$$

This, incidentally, is in good agreement with amounts known from carbide-extraction work.

Using the known carbon content of about 2·0 weight-% C in η-carbide and of 0·73% C in the total steel, the carbon content in austenite may be computed as in the range 0·3–0·4% C. This again would support alternative (*b*) above, because such a considerable austenitic-carbon content would have meant a radical break in η-carbide expansion during the $\alpha \rightarrow \gamma$ transformation had the latter caused any major denudation of carbon from the η-lattice. Thus, although the carbon in the austenite came from η-carbide, it is clear that in dissolving it, the austenite must also have taken its accompanying elements (i.e. the entire carbide) into solution.

Considering the expansion coefficient of η-carbide, it is of interest to note that, if we calculate the average $\bar{\beta}$ of the known coefficients of expansion of the elements Fe, W and C, weighted in atomic proportions as in η, Fe_4W_2C, the figure obtained is $\bar{\beta} = 9{\cdot}1 \times 10^{-6}$. Despite the obvious difference in bonding when the elements are combined, this is in close approach to the value of $9{\cdot}8 \times 10^{-6}$ as observed for η.

This may be fortuitous but the near approach nevertheless suggests that, within the lattice of the compound, the mean change of frequency of thermal vibrations with temperature remains similar to that for the component elements.

(3) Relation of Carbide and Matrix Expansion

Although the coefficients of expansion of η-carbide are somewhat lower than those of ferrite, they are still of the same order of magnitude (*see* Figs. 3 and 4). They are considerably smaller than those of austenite. With increasing temperature, the α- and η-coefficients

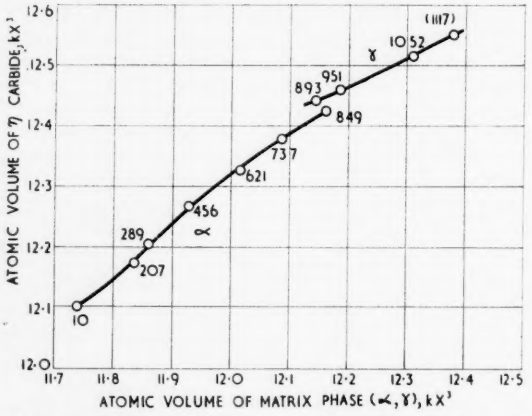


Fig. 5—Atomic volume of η-carbide against that of coexisting α- or γ-phase, at temperatures marked in figures (° C); Series *B* samples

Table IV

COEFFICIENTS OF EXPANSION OF SOME CARBIDES AND HARD METALS

Material	Temperature Range, °C	$\beta \times 10^6$	Reference
Tungsten carbide WC	20–2400†	5·2 (in 100 direction) 7·3 (in 001 direction) 5·9 mean*	Becker[18]
Tungsten carbide W_2C	20–1930†	1·2 (in 100 direction) 11·4 (in 001 direction) 4·6 mean*	Becker[18]
Titanium carbide TiC	20–600	7·3	Schwarzkopf and Kieffer[19]
Zirconium carbide ZrC	20–600	6·7	Schwarzkopf and Kieffer[19]
Tantalum carbide TaC	20–2400†	8·2	Becker[18]
Hard metals:			
WC + 6% Co	20–100	4·0	Measured by Jessop Research Dept.
	20–600	5·8	Measured by Jessop Research Dept.
	20–1000	6·5	Measured by Jessop Research Dept.
WC + 6% Co	Not stated	5	Smithells[20]
WC + 11% Co	Not stated	5.5	Smithells[20]
WC + 16% TiC + 6% Co	Not stated	6	Smithells[20]
η-carbide	10–600	10·2 (and see Fig. 3)	Present work

* **Note: For hexagonal crystals in polycrystalline aggregate, the mean coefficient is: $\beta = \frac{2}{3}\beta_\perp + \frac{1}{3}\beta_{||}$, where $\beta_\perp$ and $\beta_{||}$ are the coefficients normal and parallel to the c-axis.**

† **The large temperature range used here probably weights the β values on the high side.**

gradually diverge and, on entry into the austenitic range, the carbide expansion remains substantially unaffected by the coexistence of either α or γ.

It is instructive to plot, as in Fig. 5, a " coexistence graph " in which the atomic volumes of η-carbide are related to those of α or γ associated with it at any one temperature. We obtain two almost linear plots, of mean slopes:

$$\Delta A_\eta/\Delta A_\alpha = 0{\cdot}78 \text{ for ferrite,}$$
$$\Delta A_\eta/\Delta A_\gamma = 0{\cdot}44 \text{ for austenite.}$$

Since the ratio A_η/A_α is 0·975 (constant to within $\frac{1}{2}$%), the mean ratio of expansion coefficients $(\Delta A_\eta/\Delta A_\alpha)(A_\alpha/A_\eta)$ becomes for ferrite: $\beta_\eta/\beta_\alpha = 0{\cdot}77$.

The physical reason for this relationship will be considered in the Appendix in the light of Grüneisen's law.

(4) Relation to other Carbides

The coefficient of expansion of η-carbide is appreciably higher than that of other hard carbides, as far as available data can show. This is a fact which, in conjunction with its similarity to the expansion coefficient for iron, would appear to be of some practical interest (*see* Section 7).

Table IV gives the mean expansion coefficients for some transition-metal carbides and hard metals of the cemented-carbide type to compare with that of η. (No information seems to be available on the expansion coefficients of Fe_3C, $M_{23}C_6$, VC and other carbides prevalent in steels; their measurement ought certainly to be carried out as an important task; the X-ray method probably would again be the most suitable one.)

In Table IV, Becker's early data[18] on WC and W_2C (hexagonal) are noteworthy in showing anisotropy in the sense of increased expansion along the c-axis, namely, the direction of weaker cohesion normal to the basal planes.

It may be seen, from the cemented-carbide values, that the addition of a binder (cobalt) does not materially increase the expansion of the carbide base partly, no doubt, because it has gone into partial solution in the WC lattice. This forms the opposite extreme to the high-speed steel case, since we have:

(i) Hard metal = coherent carbide matrix with interspersed metal.
(ii) high-speed steel = coherent metal matrix with interspersed carbide.

In each case it is the matrix which mainly governs the external bulk dilatation. This theme will be followed up in Section (5).

(5) Comparison with Bulk-expansion Data on High-speed Steel

It is of interest to compare the present X-ray results with known bulk-dilatometric data on high-speed and related steels and such data are summarized in Table V. As may be seen, agreement between X-ray and bulk coefficients is very close indeed.

For a two-phase mixture, the results from the two methods are by no means bound to agree, because the individual phase expansions, obtained by X-rays. may give a resultant bulk expansion, which is either (*a*) some weighted mean between that of the two phases or (*b*) controlled by that of the dominant phase. The latter is especially likely if the major constituent forms a coherent matrix around separate cystals of the minor one (carbide), and plastic flow can occur at the boundaries.

Table V shows that, throughout, the bulk coefficients lie intermediate between the X-ray coefficients

for ferrite and η-carbide. However, they approach the ferrite value considerably more closely and in the case of Wever's data[21] we have practical coincidence from 200° C upwards (*see also* Fig. 4). The accuracy is not enough to warrant quantitative conclusions from these differences (particularly when one considers the varying sample histories) but it is clear that the ferrite matrix is the dominating factor with the carbide tending to decrease the coefficient. The Jessop dilatometic data, in fact, lie by a ratio of about 2 : 1 nearer the ferrite than the η-coefficient (e.g. β_m 20–100° C : 35%; 20–400° C : 24%; 20–600° C : 24%; 20–800° C : 33% towards η). This is in the known order of volume percentage of η-carbide in the annealed steel.

The decision between the alternatives (*a*) and (*b*) above would thus seem to be not an exclusive one and may vary with the physical condition of the sample. It may well depend on its texture, i.e. the mode of relative distribution of carbide and matrix grains; if, for example, considerable banding or preferred segregation in the rolling direction has occured, as shown schematically in Fig. 6(*a*), the direction in which the bulk dilatometry sample had been cut would be important whilst it would not affect the X-ray result. Figure 7 shows two X-ray microradiographs of a high-speed steel with pronounced directional carbide clustering: in such a case, the macro-expansion coefficient parallel to the carbide fibre would be influenced by that of the carbide more than the coefficient normal to it. Even in a case of isotropic distribution of carbide crystals within the matrix, as in Fig. 6*b*, it is still possible for the bulk coefficient to be weighted towards the carbide value to a varying extent, depending upon the degree of coagulation.

Thus the carbide expansion would exert a larger gross influence in case (I) of Fig. 6*b* than in the case (II) in which the intervening ferrite parts would permit the take-up of strains through differential expansion.

Table V also quotes some published bulk expansion coefficients for other tungsten steels; the variation for varying compositions is not very considerable. Again, a close approach to the present X-ray values may be noted.

(6) Internal Composition Changes in High-speed Steel Phases

The thermal expansion curves for the matrix and η-carbide phases are now to be considered in the light of 'physical' and 'chemical' (or solute) expansion, as defined above.

In the ferritic region, provided equilibrium has been

Table V

DATA ON BULK DILATOMETRIC EXPANSIONS FOR HIGH-SPEED AND RELATED STEELS

Steel	Temperature, °C	Coefficient of expansion ($\times 10^6$) Bulk		X-ray†	
		(Mean)		(α)	(η)
18/4/1/0·7C, annealed*	20–100	11·4		$11·6_5$,	$9·8_5$
	20–200	11·4		$12·2_0$,	$9·9_0$
	20–300	11·6		$12·4_0$,	$9·9_5$
	20–400	11·9		$12·4_5$,	$10·1_0$
	20–500	12·0		$12·5_5$,	$10·1_5$
	20–600	12·3		$12·9_5$,	$10·2_0$
	20–700	12·6		$13·4_5$,	$10·3_5$
	20–800	12·8		$13·9_5$,	$10·5_0$
18 4 1 0·7C[21]		(True) Annealed	Hardened and tempered		
	20	10·3	10·0	—	—
	100	10·9	10·7	11·8,	9·9
	200	11·7	11·5	11·8,	10·0
	300	12·5	12·2	12·5,	10·2
	400	13·1	12·8	13·2,	$10·3_5$
	500	13·6	13·4	13·9,	$10·5_5$
13·5W/4Cr/0·7C, annealed[22]	100	10·7		11·8,	9·9
	200	11·75		11·8,	10·0
	300	12·75		12·5,	10·2
Palin W Steels, condition probably as-rolled[23]		15W/0·6C	20W/0·6C		
	100	11·9	11·7	11·8,	9·9
	500	14·0	13·9	13·9,	$10·5_5$
	850	19·0	19·1	19·0,	11·1
4W/0·4C, annealed[24]	25–100	11·1		$11·6_5$,	$9·8_5$

* Kindly supplied by Mr. E. Johnson, William Jessop and Sons Ltd.
† Present X-ray results on 18/4/1 steel (for comparision)

obtained, α and η coexist essentially without atomic transfer and we are dealing with physical expansion. A very slow migration for instance of chromium into ferrite is not excluded.

In the austenitic region, the chemical expansion of the γ-matrix operates strongly as is shown by the greatly increased rate of expansion compared with pure iron. The causes are interstitial carbon and the large substitutional tungsten atoms, but also much chromium and vanadium (*see below*).

The sources of the γ-solutes are, firstly, part of the η-carbide and secondly, the minor carbides $M_{23}C_6$ and V_4C_3.

Kayser and Cohen,[25] using quantitative extraction methods, have demonstrated the increase of carbide solution in the matrix with increasing austenitizing temperatures between about 1000° and 1300° C; they have shown that the total volume percentage of carbide can be reduced by about 50% (from 28% to 15%), as between annealing and commercial hardening. The alloy-enrichment of the matrix may also be seen by the figures cited above (Section 2) which agree well with Kayser and Cohen's results.

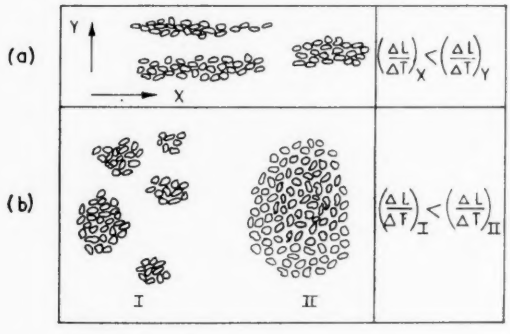
An attempt may be made to " analyse " the present expansion curves on austenite (Figs. 1 and 2) into a chemical and a physical contribution. Such an analysis would depend on the knowledge of composition variation of the matrix. Data are at present insufficient for this to be done accurately, but the following facts may be deduced at this stage.

The lattice contraction due to the $\alpha \rightarrow \gamma$ transformation is considerably smaller in high-speed steel than in pure iron, owing to the presence of alloys. The observed values are:

Volume contraction during $\alpha \rightarrow \gamma$ change of pure iron $\Delta A = 0{\cdot}130 \mathrm{kX}^3$ or $\simeq 1{\cdot}1\%$.

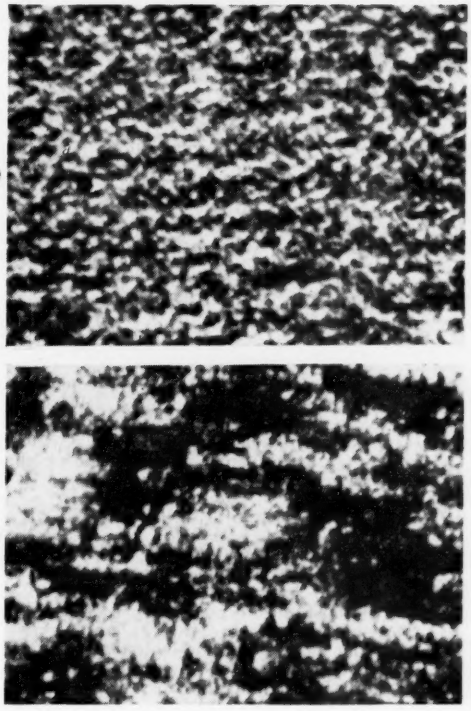
Volume contraction during $\alpha \rightarrow \gamma$ change of high-speed steel $\Delta A = 0{\cdot}035 \mathrm{kX}^3$ or $\simeq 0{\cdot}3\%$.

The transformation in the steel occurs at a somewhat lower temperature, however, than in pure iron, so that only part of this difference is chemical but for a constant temperature, namely the A_3 point in Fe (910° C), the total volume increase between γ (Fe) and γ (H.S.S.) is $\simeq 0{\cdot}110 \mathrm{kX}^3$ or $0{\cdot}9\%$, which is entirely chemical and due to the total of alloy elements $\Sigma E = \%\mathrm{C} + \%\mathrm{W} + \%\mathrm{Cr} + \%\mathrm{V}$ in solution in the γ (H.S.S.).



(*a*) Fibre aggregates.
(*b*) Non-fibre, but carbide clustering versus even dispersion.

Fig. 6—Types of carbide-grain distribution in ferrite matrix likely to affect external bulk expansion to varying degree. (See text and Fig. 7)



(*a*) Fibre-distribution (× 500)
(*b*) Fibre-distribution and clustering (× 500)

Fig. 7—X-ray microradiographs (cobalt radiation) of 18/4/1 high-speed steel, to illustrate differences liable to affect external expansion (see Fig. 6)

Use may be made of Kayser and Cohen's work (referred to above) on the carbide and matrix analyses of high-speed steels quenched from different temperatures. From their data, it can be calculated that in the matrix the total alloy content is as shown in Table VI.

Figure 8 illustrates the proportions of elements. The totals shown may be equated with ΣE above and when combined with the present X-ray data we find the following:

(i) The mean increase in alloy solution in γ per °C is $\simeq 0{\cdot}024$ % ΣE.

(ii) Between 900 and 1100° C, the increase per °C in the difference between the atomic volumes of pure γ (Fe) and γ (H.S.S.) is $\simeq 0{\cdot}00025 \mathrm{kX}^3$; the mean expansion per percentage of total alloy added is therefore $= 0{\cdot}01 \mathrm{kX}^3$.

(iii) The measured austenitic mean coefficient of expansion (by volume), which is $3\beta = (\Delta V/\Delta T)(1/V) \simeq 93 \times 10^{-6}$ per °C from the present results, can be subdivided into $3\beta_p + 3\beta_c$, the physical and chemical components. From the observed values of atomic volume expansion we have:

$$\frac{\Delta V_p}{\Delta T} = \frac{\Delta V \text{total}}{\Delta T} - \frac{\Delta V_c}{\Delta T}$$

$= 0{\cdot}00114$ (the gradient for γ as observed) $- 0{\cdot}00025$ (from item (ii) above)

$= 0{\cdot}009 \mathrm{kX}^3/°\mathrm{C}$

and division by the mean atomic volume gives: $3\beta_p \simeq 73 \times 10^{-6}$ and $3\beta_c \simeq 20 \times 10^{-6}$

Table VI

TOTAL NON-Fe CONTENT IN MATRIX OF 18/4/1 STEEL, COMPUTED FROM DATA BY KAYSER AND COHEN[25]

Heat treatment	Alloy content, wt-%				
	Total	C	W	Cr	V
As annealed	4·6	0	1·0	3·4	0·2
Quenched 1040° C	8·6	0·07	3·7	4·2	0·62
Quenched 1150° C	10·9	0·18	5·6	4·4	0·80
Quenched 1205° C	12·8	0·27	7·2	4·5	0·85
Quenched 1260° C	13·9	0·35	8·1	4·5	0·93
Quenched 1288° C	14·5	0·37	8·8	4·4	0·97
Quenched 1300° C	14·75	0·38	9·0	4·4	0·98

In linear terms, it may therefore be concluded that for austenite the observed coefficient of about 31×10^{-6} can be broken down roughly as:

$\beta = 24{\cdot}5 \times 10^{-6}$ (physical) $+ 6{\cdot}5 \times 10^{-6}$ (chemical).

(iv) We may define a mean ' coefficient of alloying expansion ' ϵ as:

$$3\epsilon = \left(\frac{\Delta V}{\Delta \Sigma E}\right)\frac{1}{V}$$

which works out as $8{\cdot}1 \times 10^{-4}$ (volume) or $2{\cdot}7 \times 10^{-4}$ (linear)/weight-% ΣE; i.e. the specific expansion through lattice solute in the austenitic range, per percentage of alloy solute, is in the order of 3μ/cm.

The obvious next step would be to try and apportion this total amongst the individual elements C, W, Cr, and V, but this is possible only very tentatively, assuming a proportional division as in Fig. 8. A conversion to atomic-% will be carried out at the same time as this gives a truer picture of the relative role of each element, particularly W and C.

Chemical expansion between 910 and 1100° C, $\Delta V = 0{\cdot}046$ kX³ total.

Through conversion from Kayser and Cohen's results:

$\Delta \Sigma E_a = 0{\cdot}53$% C $+ 0{\cdot}95$% W $+ 0{\cdot}91$% Cr $+ 0{\cdot}45$% V (in atomic %, suffix a).

The expansion due to each element then becomes:

C = 0·0085, W = 0·0156, Cr = 0·0146, V = 0·0073kX³.

Estimating that, in the austenite lattice, out of the total chemical expansion approximately

18·5% is due to gain in carbon atoms
34·0% ,, ,, ,, ,, ,, tungsten atoms
31·5% ,, ,, ,, ,, ,, chromium atoms
16·0% ,, ,, ,, ,, ,, vanadium atoms.

We also obtain, in atomic-%, the linear coefficient of alloying expansion (*see above*) as:

$$\epsilon_a \simeq 4{\cdot}4 \times 10^{-4}/\%\Sigma E_a.$$

Note regarding the $\alpha - \gamma$ transformation—One remarkable feature may be seen from Fig. 8. If the matrix-alloy-content/temperature plot for austenite is extrapolated to lower temperatures, it intersects the horizontal for the ferritic condition as annealed at a temperature (870° C) which coincides with the $\gamma \rightarrow (\gamma + \alpha)$ transformation point as determined in the present X-ray work (between 850° and 890° C) and as is otherwise known. Moreover, not only does the agreement in Fig. 8 occur by extrapolation of the *total* alloy content, but also, within the given accuracy, by separate extrapolations of the W, Cr, and V contents.

This coincidence is noted here as an empirical fact which, however, does seem significant and of considerable interest. It would indicate that, at the change-point, the face-centred and body-centred structures contain equal amounts of solute atoms so that no abrupt solution of carbide need accompany transformation. This will be further discussed in Part II.

(7) Note on a Practical Aspect: Sintered η-carbide

In a separate investigation[26] it had been proposed to utilize η-carbide powders extracted from a high-speed (or related) steel, when pressed and sintered, as an inexpensive and attractive hard metal. The possibilities and limitations of this plan were examined in some detail. One significant property is that, in contrast to the WC, TiC, etc. of conventional hard metals, the coefficient of expansion of η-carbide is similar to that of α-iron and of ferrite in steel and the present work may be regarded as providing the proof (and the qualifications).

In technical practice, this would suggest the use of η-carbide sinters in any applications where hardness or wear-resistance together with thermal expansion comparable with that of iron or steel are desired properties. Examples of such applications are precision gauge blocks, hard inserts in steels, multiple components laminated with more ductile material, or hot-die and mould linings. Also, if carbide units are diffusion-sintered on to steel, for instance high-speed steel itself (which can readily be done with η), it would follow that the internal stresses set up at carbide/ferrite crystal boundaries become a minimum, owing to the small differential expansion.

The differential expansion between η and austenite, on the other hand, would seem too large for these points to apply.

SUMMARY AND CONCLUSIONS

(1) It is possible, by the X-ray high-temperature method, to measure the separate expansions of two phases in an alloy, a principle here exemplified by application to the matrix and η-carbide phases of a high-speed steel.

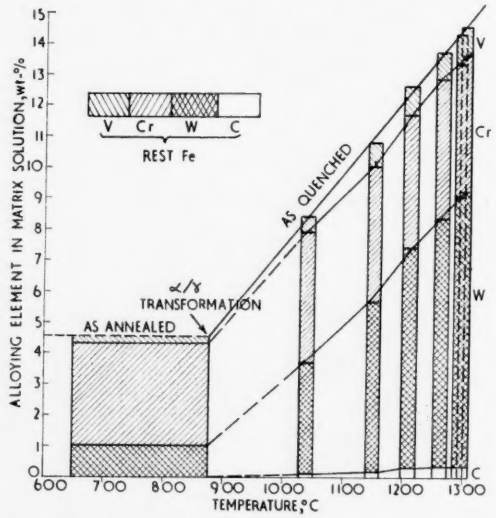


Fig. 8—Solute-content in austenite and ferrite matrix solution for an 18/4/1 steel quenched from various temperatures and as-annealed. Derived from Kayser and Cohen's data.[25]

(2) For ferrite, the lattice expansion is less than for pure α-iron, typical observed coefficients being:

	α - *H.S.S.*	α - *Fe*	
10–200° C	12·2	12·60	$\times 10^{-6}$ per ° C
10–800° C	13·95	14·65	

For austenite, the lattice expansion is greater than for pure γ-iron due to progressive alloy solid solution, typical coefficients being:

	γ-*H.S.S.*	γ-*Fe*
950–1050° C	32·2	$23{\cdot}3 \times 10^{-6}$ per ° C

(3) The $(\alpha + \gamma)/\gamma$ transformation temperature was found in the region of 870° C.

The volume contraction occurring during the $\alpha \rightarrow \gamma$ change is appreciably less in high-speed steel than in pure iron.

(4) The carbide expansion is practically linear with a true coefficient between 10 and 11×10^{-6} up to 850° C; at higher temperatures the rise is slightly more rapid.

A small step *P*, which occurs at about 900° C and can be related to a much more pronounced one *P'* found for the less equilibrated series, is probably due to certain metastable reactions. However, in equilibrium, the fact whether α or γ is the associated phase makes remarkably little difference to the expansion of η-carbide.

(5) The mechanism of η-carbide solution in austenite is substantially one of reduction in total quantity rather than carbide-composition change, though to a small extent the latter also applies.

(6) The coefficient of expansion of η-carbide is close to that of α-iron or steel ferrites, in contrast to the coefficients of other known hard carbides (WC, TiC, etc.) which are appreciably lower. This property of η suggests that it might be utilized in certain practical applications in which similarity in expansion of a steel and a hard metal is desirable; this applies particularly to sintered η-carbide compacts made from the extracted powder.

(7) The ratios of the atomic volumes of the carbide to those of the matrix-phase coexisting with it at given temperatures are:

$$\Delta A_\eta/\Delta A_\alpha = 0{\cdot}78;\ \Delta A_\eta/\Delta A_\gamma = 0{\cdot}44.$$

This can be explained qualitatively (*see* the Appendix) by the theory of thermal expansion. Assuming Grüneisen's law, specific heat and density (known from η-carbide sinters) can be correlated with expansion coefficient. The computed compressibility for η-carbide is approximately $0{\cdot}25$–$0{\cdot}3 \times 10^{-12}$ dyn/cm^2, i.e. rather less than half that of iron.

(8) The observed X-ray expansions agree well with those known from bulk dilatometry for high-speed steels, the coefficients for ferrite and η-carbide closely flanking the bulk values. The phase principally controlling the macro-expansion coefficient is the matrix but a lowering effect, due to the carbide, is evident; its quantitative effect may be greater or smaller, depending on the mode of phase aggregation.

(9) The lattice expansion of the matrix phases can be divided into a 'physical' and 'chemical' contribution ($\beta = \beta_p + \beta_c$), the latter being due to the increase of solute atoms alone. For ferrite in equilibrium, it is substantially $\beta = \beta_p$, i.e. no $\eta \leftrightarrow \alpha$ interaction occurs. For austenite (and martensite) β_c becomes considerable; the composition of the austenite matrix can be calculated from Kayser and Cohen's analytical data on quenched high-speed steels and, combining this with the present expansion results, a number of data may be deduced. These are summarized in Table VII (valid for temperature range 900–1100° C).

Table VII
SUMMARY OF DATA

Quantity	Computed Value (approximate)
Increase of total alloy solution in austenite (C + W + Cr + V), per ° C	0·024 weight-%
Lattice expansion of austenite as from pure γ-iron, per ° C (by volume)	0·00025kX^3
Mean volume expansion per % total alloy solute added	0·010kX^3/weight-% solute 0·016kX^3/atomic-% solute
Volume coefficient of expansion for austenite:	
total	$3\beta \simeq 93 \times 10^{-6}$
physical	$3\beta \simeq 73 \times 10^{-6}$
chemical	$3\beta_c$ — 20 $\times 10^{-6}$
'Coefficient of alloying expansion' (volume) per percentage total alloy solute added	3ϵ — $8{\cdot}1 \times 10^{-4}$/wt-% solute $3\epsilon a \simeq 13{\cdot}2 \times 10^{-4}$/atomic-% solute
Total chemical expansion of austenite between 910 and 1100° C of which* gain in:	0·046kX^3
C-atoms contributes:	18·5%
W " " :	34%
Cr " " :	31·5%
V " " :	16%

* Tentative, assuming proportional division as in Fig. 8.

(10) The solute-content/temperature graph for the austenitized steel matrix is found to intersect the solute-content of ferrite in the annealed steel at a temperature coinciding with the $(\gamma/\gamma + \alpha)$ transformation temperature as located here by X-rays; this, in conjunction with the continuous change in expansion coefficient during the transformation, gives certain evidence regarding martensite-type transformations, to be amplified in Part II.

Acknowledgments

The writer wishes to express his thanks to Mr. D. A. Oliver, Director of Research, The B.S.A. Group of Companies, for his interest and encouragement, to Mr. J. A. Brand, for valuable assistance, and to Mr. E. Johnson, for comments and information on some of the bulk dilatometric data.

APPENDIX

Considerations on the Relative Expansion Coefficients of Ferrite and η-Carbide (Cf. Section (3) of Discussion)

The physical reason for the relation of matrix and carbide expansions observed will here be briefly considered from the aspect of basic theory, as this brings out some points of considerable interest.

Table VIII

COEFFICIENTS OF EXPANSION AND SPECIFIC HEAT OF FERRITE AND η-CARBIDE

	Room-temperature		600° C	
	$\beta \times 10^6$ (per ° C)	c_v (cal per ° C per g)	$\beta \times 10^6$	c_v
Ferrite	**11·65**	**0·106***	**14·55**	**0·132**
η-carbide	**9·85**	**0·062**	**11·0**	**0·069**

*** Pure Fe value used, assumed the same as for ferrite in high-speed steel.**

From the equation of state, we have:

$$\text{volume coefficient, } 3\beta = \frac{-\chi_0 c_v}{V_0}\frac{(\mathrm{d}\log \nu)}{(\mathrm{d}\log V)} \qquad (1)$$

where χ_0 = compressibility at absolute zero,
V_0 = volume per gramme at absolute zero,
V = volume at temperature,
c_v = specific heat at constant volume,
ν = frequency at atomic vibrations.

This expresses that a high coefficient of expansion is associated with the greater increase in thermal-vibration frequency per unit-volume change, and with a weakening in bond strength. It is consistent that the hard transition-metal carbides, like WC, with their increased cohesive energy and lower compressibility should, broadly speaking, have the lower expansion coefficients. (In a simple model we might visualize the interstitial carbon atom as not only increasing the strength of bonding, but also acting as a damping agent to the vibrations of the metal atoms).

For η-carbide, this would mean a lower expansion coefficient than for iron, as in fact is the case, but to a lesser degree than for the hard carbides of simpler structure, e.g. WC, TiC. As an empirical guide, melting point is a preferable criterion to hardness (short of heat of sublimation data, which are lacking). The melting point for η-carbide is low compared with other hard carbides (namely, approximately 1650° C), and not so very much above that of iron (1537° C) or of the ferrite matrix of the high-speed steel. Similarly, the decrease in β is only moderate.

In equation (1), the term $\dfrac{d\,(\log \nu)}{d(\log V)} = \dfrac{-38\,V_0}{c_v \chi_0} = \Gamma$ is Grüneisen's number,* known to be nearly constant, namely, between 1 and 3 for all substances. That the coefficient of expansion should be proportional to specific heat, for a given substance, is a consequence of Grüneisen's Law.

Applied to high-speed steel it would be possible, assuming the validity of this law, to compute the specific heat of η-carbide, if its compressibility were known, and vice versa. Both the density ρ and the specific heat c_v of η-carbide have been measured (on sinter compacts of the extracted carbide, in separate work). They are $\rho = 11{\cdot}95$ and $c_v = 0{\cdot}062$ cal per ° C per g at room temperature.

Using Grüneisen's law, we obtain Table VIII from the observed values stated:

Using the atomic specific heats, we find:

$$\frac{(\beta/C_v)_\eta}{(\beta/C_v)_{\mathrm{Fe}}} = \frac{(\chi_0/V_0)_\eta}{(\chi_0/V_0)_{\mathrm{Fe}}}\,G,$$

where G (approximately constant) = ratio of Grüneisen numbers $\Gamma_\eta/\Gamma_{\mathrm{Fe}}$.

Comparing η-carbide with the ferrite with which it coexists at room temperature we may attempt to estimate the order of magnitude of *compressibility* for η using the known densities.

Since density $\rho_\eta = 11{\cdot}95$, $\rho_{\mathrm{Fe}} = 7{\cdot}85$, and compressibility $\chi_0(\mathrm{Fe}) = 0{\cdot}6 \times 10^{-12}$ dyn/cm² (*see* Mott and Jones[27]) we obtain $3(\beta/C_v)_\eta \simeq 5{\cdot}52 \times 10^{-6}$, $3(\beta/C_v)_{\mathrm{Fe}} \simeq 5{\cdot}91 \times 10^{-6}$ (volume coefficients) where $\chi_0(\eta)/\chi_0(\mathrm{Fe}) \simeq 0{\cdot}62$ or $\chi_0(\eta) \simeq 0{\cdot}37 \times 10^{-12}$ dyn/cm².

This applies if $G \simeq 1$, while the Grüneisen constants for Fe and η may differ. We know $\Gamma_{\mathrm{Fe}} = 1{\cdot}6$[27] but Γ_η is likely to be of the order of 2·5, so that we can estimate the compressibility of η-carbide to be in the range $\chi \simeq 0{\cdot}25$–$0{\cdot}3 \times 10^{-12}$ dyn/cm² or rather less than half that of ferrite.

* Symbol Γ used instead of the usual γ, to avoid confusion with present γ-phase.

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Part II—TRANSFORMATIONS DURING TEMPERING AND SOME CONSIDERATIONS ON MARTENSITE-TYPE REACTIONS

SYNOPSIS

The phase transformations during the tempering of a quenched 18/4/1 type high-speed steel are followed at three tempering temperatures, 500°, 550°, and 600° C, by the high-temperature X-ray method, and the progress of isothermal austenite → martensite → ferrite decomposition directly observed; the differences in rate and manner of this reaction are discussed, and the meaning of the expansion coefficients of the temper-phases considered from the viewpoint of the preceding lattice expansion study (Part I).

Some general aspects of high-speed steel kinetics are discussed, and present indications strongly support the idea that during the γ/α transformation at temperatures around and just above the Ae_1 point, a *continuous* composition change occurs between the two phases. The concept of a "high-temperature martensite," isocompositional with austenite, thus arises, and there is a degree of indeterminacy in equilibrium at temperatures of the eutectoid and pro-eutectoid regions; in high-speed steel conditions favour this, and permit ready coherency of the two phases between and even within grains. Recent theories on bainite would well support this conception.

Pseudo-equilibrium effects in the stable (ferrite + η) region are noted, including the occurrence of an 'imperfect' ferrite in the annealed steel. 1336

IN PART I OF THIS PAPER the lattice expansions of matrix phase and of carbide in a high-speed steel have been studied under equilibrium conditions. In contrast, Part II will deal with the isothermal transformation of unstable austenite on tempering, as followed by X-rays at tempering temperature, so that in a sense we have in these two parts illustrations of two opposite methodical approaches to high-speed steel by the high-temperature X-ray technique.

The well established heat-treatment normally given to high-speed steel of 18/4/1 type is to harden by oil-quenching from about 1300° C, and to temper at about 550° C; but this procedure has been elaborated in extensive work,[1-3] which has led for instance to the double and multiple tempering methods.

The breakdown of retained austenite after quenching results, on tempering, in the precipitation of temper-martensite and carbide, a reaction which is mainly responsible for secondary hardening and other effects characteristic of high-speed steel. During the 550° C temper also the 'quench'-martensite, i.e. the immediate diffusionless product of lattice-shear of austenite during the preceding quench, is partly broken down towards the ferritic state; however the detailed mechanism of this complex reaction is still far from solved.

The present investigation had the limited object of observing the isothermal $\gamma \rightarrow \alpha$ transformation in the as-hardened steel during tempering in the X-ray camera. This technique is believed to be generally well suited for investigations attempting to simulate heat-treatments of steels, provided however the experimental limitations are realized, for instance in sample size, rates of cooling, and exposure times.

EXPERIMENTAL

The material used was the same 18/4/1 type high-speed steel, as used in the work described in the previous paper. The small rod-shaped X-ray samples were first austenitized by water-quenching in evacuated silica tubes from 1300° C after 3 min holding. After initial photographs at room-temperature in the as-quenched condition, the samples were "tempered" in the high-temperature camera *in vacuo* at three constant temperatures: 500°, 550°, or 600° C, X-ray exposures being taken at various times, as shown in Table I.

RESULTS AND DISCUSSION

The constitutional results are summarized in Table I.

The only phases observed were γ, α, and η-carbide, where α covers both martensite (α_M) and ferrite (α_F); the chief distinction between the latter is that of line-width, which can be extreme for the α_M higher angle reflections. The proportions of the three phases could be estimated to the figures shown in Table I; the ratio of η-carbide to matrix increases from approximately 1 : 5 as quenched, to 1 : 2 as annealed (consistent with quantitative extraction-work and the results of Part I), and successive tempers give intermediate amounts. The progressive decomposition of austenite is also shown in Table I.

It may be seen that at 600° C the breakdown of γ proceeded with fair rapidity; it had reached about 90% after 6 h, and completion after 24 h (probably earlier). The quench-martensite showed extreme line-broadening, but during the first temper-exposure of 1½ h the ferrite condition was almost attained. However, as long as any retained austenite persists, and for a period after that, a second α-portion also persists, which is martensitic and represents the immediate product of transformation in contrast to the ferrite also present. This is visible on the X-ray photograph in the form of broad 'ghost' reflections, upon which the sharp lines of the advanced ferrite superimpose. A small amount of this background martensite appears throughout the 600° C temper, even after 24 h.

The tetragonal distortion of martensite is very small in high-speed steel; where detectable the axial ratio was $\simeq 1{\cdot}005$, but in most cases any tetragonal line-split was hidden by the line-broadening.

A very slight lattice contraction of ferrite is observed with time at 600° C due to loss of residual solute content which, as shown in Part I, would be mainly carbon and tungsten, chromium largely remaining in stable solution.*

The sequence of 600° → 550° → 500° C tempers shows the progressively slower rate of decomposition, in all three respects: decrease of γ, line-sharpening of α,

* An incidental point in X-ray technique is illustrated in Table I, namely an advantage of the counter-diffractometer method[4] over the photographic in studies of continuous phase-transformations: in the first (1½ h) exposure at 600° C the lack of resolution of events in time becomes noticeable.

Table I

PROGRESS OF AUSTENITE DECOMPOSITION ON TEMPERING IN THE X-RAY CAMERA
Material: 18/4/1 High-Speed Steel, Sample Water-Quenched from 1300° C

Times of Tempering, h	Constitution*				Remarks†
	Phases	Estimated Amounts, %			
		α	γ	η	
600° C					
0 (as quenched)	$\alpha_M + \gamma + \eta$	55	30	15	*D*+ Rapid sharpening and slight lattice-contraction.
0 to 1½	$\alpha_M + \gamma + \eta$	70	12	18	*S*−
2 to 3¾	$\alpha_F + \gamma + \eta$	70	7	23	*S*− Some martensitic background
4 to 6	$\alpha_F + (\gamma) + \eta$	70	4	27	*S* persists throughout, superimposing on α_F.
22 to 24	$\alpha_F + \eta$	70	—	30	*S*
550° C					
0	$\alpha_M + \gamma + \eta$	55	30	15	*D*+ Initial temper-sharpening, then practically no change.
0 to 2	$\alpha_M + \gamma + \eta$	58	25	17	*D*−
3 to 5	$\alpha_M + \gamma + \eta$	63	15	22	*D*−
7½ to 9½	$\alpha_M + \gamma + \eta$	65	10	25	*D*− No visible lattice-parameter change.
30 to 32	$\alpha_M + \eta$	70	—	30	*D*−
500° C					
0	$\alpha_M + \gamma + \eta$	55	30	15	*D*+
0 to 2	$\alpha_M + \gamma + \eta$	57	27	16	*D*
2½ to 4½	$\alpha_M + \gamma + \eta$	58	25	17	*D* Initial temper-sharpening, then no change.
4½ to 6½	$\alpha_M + \gamma + \eta$	61	17	22	*D*
8 to 10	$\alpha_M + \gamma + \eta$	63	12	25	*D*
10 to 12	$\alpha_M + \gamma + \eta$	64	10	26	*D*
12 to 14	$\alpha_M + \gamma + \eta$	66	8	26	*D* No visible lattice-parameter change.
31 to 33	$\alpha_M + (\gamma) + \eta$	67	6	27	*D*
188 to 190	$\alpha_M + (\gamma) + \eta$	69	4	28	*D*

* These figures are intended as an approximate guide only.
The relative changes are believed to be correct, while the absolute values may be subject to an error of up to ± 10% of the figure given.
† On martensite line-width and lattice-dimensions; see text.

KEY: α_M = Martensite; α_F = Ferrite; γ = Austenite; η = η-carbide.

Line-widths: *D*+ = Extreme martensitic diffuseness *D* = Diffuse *D*− = Moderately diffuse *S*− = Almost sharp *S* = Sharp

and increase of η. Thus at 500° C transformation is so sluggish that some austenite still persists after 190 h, and the martensitic nature is maintained throughout; except initially from the as-quenched state, no further line-sharpening occurs with time. It is thus of considerable interest that the rate of decomposition of austenite is far greater than that of martensite, the difference being the larger at the lower temperature. The η-carbide proportion gradually increases as decomposition advances, and it appears that, at this low temperature, diffusion proceeds heterogeneously and the martensitic state persists throughout, until the austenite-breakdown is nearly complete.

It is well established that, on tempering hardened 18/4/1 steel at about 550° C, maximum secondary hardness is attained after 1 to 2 h; it seems at present that this is clearly related to the stage of breakdown of the main portion of austenite, and to maximum martensite formation in good agreement with theory. The results also conform with the knowledge that the optimum tempering temperature is around 550° C, if we interpret ' optimum ' as the lowest temperature of disappearance of austenite (within a reasonable time). At 500° C austenite will still transform, albeit very slowly, but if very extended tempering periods were practicable (say 100–300 h, rather than the normal commercial times of the order of 1–5 h) equivalent or better properties would perhaps be produced, for example the martensite may be of higher hardness.

The double or multiple tempering techniques are known to produce increased hardness compared with equal total times of single tempers. The question whether austenite decomposition takes place (1) athermally on cooling *from*, rather than (2) isothermally on holding *at*, the tempering temperature, did not form part of the present investigation; but at least it can be said that case (2) certainly applies, and the question is to what extent athermal decomposition applies as well. That it does, seems certain from much evidence on the fundamental nature of martensite, [3, 5–9] but the degree of decomposition will depend on the time-temperature cooling curve below 550° C, which indicates whether the structural changes would occur by nucleation-and-growth or martensitically by lattice-shear. Besides the γ-shear transformation other factors are involved such as the stage the $\alpha_M \rightarrow \alpha_F$ reaction has progressed to during tempering by the time of intermediate cooling, and the state of precipitation and particle-size of any secondary carbide formed ($M_{23}C_6$, VC).

From early wotk[10, 11] on the dilatation of quenched high-speed steel it was concluded that retained austenite would not transform isothermally at

tempering temperature, but only on cooling from it. The present results on the other hand give direct evidence of isothermal transformation.

LATTICE EXPANSION OF NON-EQUILIBRIUM PHASES

Viewed from the high-speed steel expansion results reported in Part I, the present tempering transformations lead to some interesting conclusions.

Austenite: The expansions of *stable* austenite, observed in the γ-region at high temperatures, would suggest that the as-quenched steel with its *retained* austenite should have a higher coefficient of expansion than the tempered one. This is known to be the case from published bulk-dilatometric work on hardened and on tempered high-speed steels.[11–15] As austenite decomposes, expansion occurs in bulk, but the coefficient of expansion decreases, consistent with the inherently larger atomic volume but smaller expansion coefficient of martensite and ferrite as well as of η-carbide. Typical figures[13] are: as-quenched (austenite) $16{\cdot}9 \times 10^{-6}$; as-tempered (ferrite) $11{\cdot}4 \times 10^{-6}$ per ° C at room temperature.

With due modifications, the outcome of the equilibrium expansion curves of Part I may be related to the metastable case of the tempered steel. The coefficient β for stable austenite was subdivided into a 'physical' and a 'chemical' component, $\beta = \beta_p + \beta_c$; in retained austenite there is a considerable supercooled solute content (C, W, Cr, V), so that the reverse process operates in its decomposition and at any one stage the new value would be (approximately) $\beta = \beta_p - \beta_c$. If, however, retained austenite is regarded as 'ideally quenched,' and of unaltered composition while it persists (the burden of decomposition resting with the product martensite after its diffusionless formation), β_c can be neglected so that for any thermal expansion of retained austenite, $\beta = \beta_p$. The actual values would of course differ from the coefficients obtained before ($\beta \simeq 31$, $\beta_p \simeq 24{\cdot}5 \times 10^{-6}$), owing to the much lower temperatures concerned in retained than in stable austenite.

Martensite: The coefficient of expansion of martensite is a quantity complex in nature for three reasons. Firstly, on initial formation, the expansion will correspond to that of a pure $\gamma \rightarrow \alpha$ shear transformation, during which all solute atoms are retained. On tempering, the β_c-term becomes prominent, the more so for the higher tempering temperature, whilst the terminal value should be that of pure ferrite; thus the expansion coefficients begin and terminate as physical ones (namely β_p values for martensite isocompositional with austenite, and for pure ferrite), β_c superimposing at intermediate stages. Secondly, owing to the inherent structure of martensite (microheterogeneous in composition and highly strained), its expansion coefficient would represent, at any one stage, not a single value, but a 'band,' although statistically a single external coefficient would still arise. Thirdly, the unstable presence of the carbon atoms in the lattice at some edges and base-centres of the body-centred tetragonal unit cells, and the displacement of metal atoms from their normal positions, would modify the cohesive energy and thermal vibration amplitudes, and therefore the expansion coefficients.

Thus the coefficient for martensite is, in principle, an ill-defined quantity, in contrast to that for austenite and ferrite. The experimental determination of thermal expansion of martensite by X-rays is almost impossible, since the characteristic line-broadening vitiates any accuracy; this may be considered as an observational counterpart of the same fundamental indeterminacy.

SOME ASPECTS OF HIGH-SPEED STEEL KINETICS

(*a*) Composition Continuity during $\alpha \longleftrightarrow \gamma$ Change

It has already been observed in Part I that, even for apparent attainment of equilibrium, minor but significant structural differences are liable to occur. The two expansion-series *A* and *B* (which differed in the time of pre-anneal at 700° C) provided a case in point; the small spacing divergence occurring between them at the lower end of the austenite region for instance would suggest a slow dissolution of carbide in γ. Furthermore, it was noted that the curves of expansion coefficients for γ and α are joined in an apparently continuous curve (though of steep gradient). Lastly, there was the phenomenon (Fig. 8, Part I) of coincidence of the $\gamma \rightarrow \alpha$ change-point with the intersection of γ- and α-matrix compositions.

All this strongly indicates that the degree of solution changes continuously at and near the change-point, i.e. that the transformation occurs by a process in which initially the face-centred and the body-centred cubic lattices are of equal composition. The eutectoid reaction $\alpha + \eta \longleftrightarrow \gamma + \alpha + \eta$ of course demands an eventful change in composition between the two matrix-phases, but the rate of transformation by lattice-shear is much more rapid than that of diffusion, increasingly so on approaching the point of zero free energy difference (Ae_1 temperature).

Under conditions of cooling from austenite an interesting consequence arises in that the diffusionless formation of martensite might be considered as the kinetic complement to this process; unless the rate of cooling is sufficiently slow, carbon and metal-solute atoms will in that case be retained in supersaturation by the body-centred daughter-lattice. Thus it is possible to introduce a term 'high-temperature martensite,' this 'martensite' existing for a limited time in the region of the $\gamma \rightarrow \gamma + \alpha \rightarrow \alpha$ transformation temperatures (870° to 790° C for 18/4/1); such a 'martensite' would be consistent with the effect of a continuous transition both in composition and in expansion coefficient.

It is suggested that *at* the transformation temperature, even in true equilibrium, the W, Cr, and V solute contents of ferrite and austenite do not in fact differ appreciably, only the interstitial carbon content.

At this point, a brief consideration of the phase-diagram basic to high-speed steels proves useful, because the $\alpha + \gamma + \eta$ field occurring in it might here gain a new interpretation. Figure 1 shows Murakami and Hatta's[16] diagram for 18/4/1 type steel (recently supported by Kuo[17]), which represents a section through the Fe–Cr–W–C system. For the carbon content indicated one obtains, with increasing

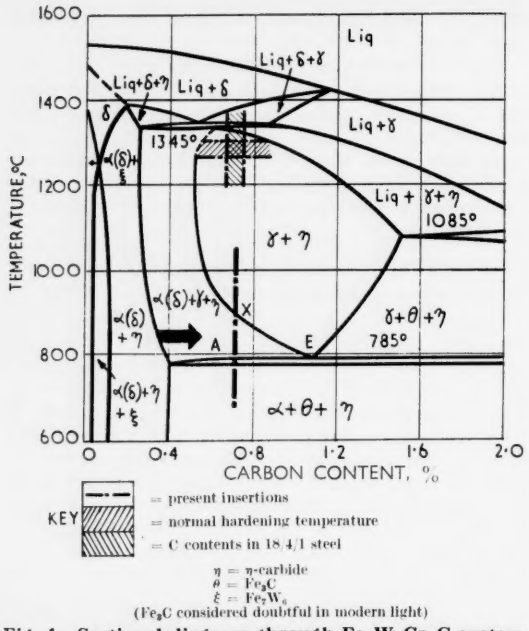

Fig. 1—Sectional diagram through Fe–W–Cr–C system at 18% W, 4% Cr, according to Murakami and Hatta,[16] as modified by Kuo[17] (broken line)

temperature, the $\alpha \rightarrow \alpha + \gamma \rightarrow \gamma$ sequence as at present observed. The agreement of the $\gamma/(\gamma + \alpha)$ boundary (proeutectoid point X, about 870° C) with that found by X-rays in Part I and with that of the γ/α composition intersection, may incidentally be affirmed in Fig. 1.*

In the above light, the continuous change between the α- and γ-phases may be interpreted by assuming that the boundary XE in Fig. 1 can sensitively adopt a sequence of metastable positions, depending on the detailed heating and cooling history around this temperature region. The above 'high-temperature martensite' would thus be a highly labile α, which can readily shear back and forth to an austenite of similar composition; this in turn might be termed a 'high-temperature retained austenite.' The region XEA is thus so extremely favourable to thermal hysteresis effects that it is even arguable that true equilibrium here cannot be defined or distinguished from a metastable one. Considering the complex nucleation processes involved in high-speed steel, with at least three potential carbides in incipient precipitation, this effect of 'self-balancing' α and γ would be well understandable.

From the phase-rule viewpoint, this indeterminate transformation may be expressed by the free-energy/temperature curves for the two phases widening into a band in the neighbourhood of the point of intersection, and conditions in region XEA are favourable for this to occur. In Fig. 1 it may formally be expressed by an approach of the $\alpha + \eta$ and $\gamma + \eta$ phase-boundaries, as indicated by the bold arrow.

* There are however some inconsistent features in Murakami and Hatta's diagram, such as the presence of phase 'θ' (cementite), and the omission of $M_{23}C_6$; the thorough re-determination of this complex system would certainly be desirable.

Near the transformation point, fluctuations in α/γ ratio are very likely to occur in a single specimen from grain to grain through the fact that compressive stresses produced by the transformation in one crystallite may constrain its neighbours and retard their transformation. Even within the frame of a single crystal, under conditions of vanishing free energy difference as at present, it is envisaged that incipient face-centred and body-centred configurations can coexist at different lattice regions, depending on distance from grain-boundary and factors such as lattice imperfections and local occlusions of impurity atoms.

Dislocations in particular are well capable of forming boundaries between sub-domains of α within a γ-crystal and conversely, and their movement through the crystal may cause a given group of atoms readily to transfer to and from parent and product lattice. For undercooled austenite the dislocation movement would of course act in one direction only, so that the application of an external alternating stress would probably tend to accelerate the breakdown of γ; such a strain-induced $\gamma \rightarrow \alpha$ transformation, produced for instance by mechanical work during tempering the steel or by the application of a magnetic field during treatment, would, it seems, offer a useful field for further research (it is already known that undercooled austenite can transform to martensite above M_s when plastically deformed). Hybrid single crystals of austenite and martensite existing at and just above the Ae_1 temperature are, it is suggested, a fundamentally important feature in many aspects of high-speed steel theory and behaviour.

The phenomena referred to above occur around the $\gamma \longleftrightarrow \alpha$ equilibrium transformation range, but can be inhibited by quenching, and at lower temperatures are then translated into the metastable reactions of tempering as discussed in the following.

(*b*) Imperfect Equilibria at Lower Temperatures and Effect of Preceding Heat-Treatment

At temperatures between the eutectoid and M_s (as on tempering the quenched steel) the above effects can be reproduced on a different time and energy scale; the tempering results obtained here might be read as fundamentally similar in cause, except that now the isothermal reaction proceeds irreversibly towards α. It is clear that the diffusionless martensitic transformation and the nucleation-and-growth mechanisms are intimately coupled, during the whole $\gamma \rightarrow$ martensite $\rightarrow$ ferrite reaction series, and reference need here only be made to the many studies on S-curves, and pearlite/bainite formation during the eutectoid reaction; metallographically the above X-ray observations for the 600° and the 500° C time-sequence (*see* Table I) will represent two typical cases, near-opposite extremes.

Even in the fully ferritic condition, traces of non-equilibrium solutes are still able to persist in the α-lattice; although the phase is 'ferrite' in the accepted sense, such imperfections might be considered as a last residue from the martensitic reaction. The influence of pre-heat-treatment on a sample in this ostensibly equilibrated (ferrite + η) condition is readily revealed by small variations in lattice-spacing.

Examples for the present high-speed steel are given in Table II.

In two other instances pre-anneals under different conditions gave the pairs: 2·8643/11·073kX and 2·8630/11·075kX. Significant differences therefore can occur amongst pure ferrites due to slow diffusion of trace alloy elements in solution. (The α- and η-spacings are not necessarily interdependent variables because either phase can gain metastable solute from the minor carbide $M_{23}C_6$ as well as from martensite.)

Such pseudo-equilibrium fluctuations within ferrite may, it is suggested, have an appreciable bearing on practical tempering behaviour and on properties. Discrepancies and apparent lack of reproducibility in the bulk dilatation of commercially tempered high-speed steels have often been noted in the literature; this may readily be explained on the present evidence not only by an ill-defined stage in the austenite → martensite → ferrite breakdown but also by the present near-equilibrium changes. As already discussed in Part I, such deviations could give an 'apparent' rather than real coefficient of expansion; this is better revealed by the X-ray method than by bulk dilatometry because in bulk any concurrent γ/α transformation may obscure the component causes of individual phase expansions.

A practical point is that this type of variation in relative expansion of ferrite and carbide could offer the perhaps useful possibility of a degree of deliberate adaptation between volume-changes of the two phases in the high-speed steel mass. As shown in Part I the difference in α- and η-expansion at different temperatures can be controlled within limits by preceding heat-treatment; it may well help in a given application thus to minimize the relative expansion in this way in order to minimize internal stresses further than the closeness of η-carbide and ferrite coefficients already permits.

(c) Lattice Considerations

It may be shown that, in 18/4/1 type high-speed steel, the compositional conditions are favourable for the coherent coexistence of the γ- and α-lattices.

From known analysis results on electrolytic extracts,[18, 19] the atomic composition of the *matrix* can be calculated as in Table III, for the extremes of full hardening and full annealing, and for one intermediate case.

Therefore the whole reaction mechanism during heat-treatment takes place within the narrow confines set by compositions (1) and (2). Structurally, this means that, as hardened, there is in the f.c.c. γ-lattice only 1 carbon atom per about 20 unit cells, and the same in the b.c. tetragonal quench-martensite associated with it. Thus the lattice population of carbon-atoms is relatively small (smaller than in equivalent plain carbon steels, where cementite, unlike η-carbide, entirely dissolves in γ).

It seems, therefore, that in high-speed steel the martensite lattice-strain should not primarily be attributed to the interstitial carbon. It is more likely that short-range displacement and ordering of metal atoms, W, Cr, V and Fe, already occurs within the parent lattice, showing a tendency towards octahedral configurations around the carbons (as in η-carbide), and that this major metal-atom displacement comprising a group of unit cells is a principal feature of the present martensite, and indirectly contributes to X-ray line-broadening and hardness. (We assume here that longer-range heterogeneity in lattice-composition has not yet become operative, as it will on subsequent tempering, and that we are dealing with the martensite 'islands' coherent with the γ-parent lattice, as discussed above.)

In addition we have the normal Bain-type shear strain, such as occurs in Fe–Ni martensites, which is independent of the interstitial atom. One might then regard high-speed steel martensite as a hybrid between these two modes of martensite: that derived from a metal only and from a carbon-nucleation transformation. The situation is further complicated by there being not one, but three potential carbide-precipitates (η, $M_{23}C_6$ and VC), a factor which would also contribute to the retainability of austenite by tending to inhibit precipitation.[19]

A further factor is that in high-speed steel the volume-changes associated with the $\gamma \rightarrow \alpha$ transformation are relatively small, as shown by the lattice dimensions observed in Part I, so that such martensitic domains could be tolerated fairly readily within the austenite crystal, without strains exceeding those required to produce nucleation; homogeneous change of shape associated with martensite could probably be similarly accommodated within small domains, without disrupting the coherency of the lattice. (See also Bilby and Christian's review on martensitic transformations.[6])

All these points would contribute towards a sluggish transformation kinetics in high-speed steel, as indeed observed (e.g. temper-series at 500° C). It might be said that the unique hardening characteristics of

Table II

SOME RESIDUAL CHANGES IN LATTICE SPACINGS OF FERRITE AND CARBIDE

Time at 700° C (Initial condition as-annealed: ferrite + η)	Lattice-spacing, kX, at room temperature	
	a	η
½ hour	2·8638	$11·063_6$
2 hours	2·8637	$11·063_6$
7 days	2·8633	11·064
28 days	2·8632	11·066

Table III

COMPOSITION AND CONSTITUTION OF 18/4/1 STEEL MATRIX AFTER THREE TREATMENTS

	Matrix Composition (atomic-%)					Matrix Phases
	Fe	W	Cr	V	C	
(1) Quenched 1290° C	89	2·7	4·9	1·2	2·4	Retained austenite + martensite
(2) Fully annealed	96	0·5	3·25	0·25	0	Ferrite
(3) Quenched + tempered 550° C, 2 hours	92	0·4	3·9	0·8	1·4	Martensite + retained austenite

high-speed steels, in distinction to other tool steels, are due largely to the fact, that, for the particular ratio of atoms Fe : W : Cr : V : C, a favourable balance of all these structural conditions happens to be just attainable.

The above has a direct bearing on the character of bainite. In bainite, we should probably still see essentially a product of martensitic lattice-shear, but at an early stage of nucleation-and-growth of ' ferrite ' plus carbide, a small degree of eutectoid diffusion having taken place, which increases from zero on passing from lower- to upper-bainite. The ' ferrite ' constituent should, in the light of this paper, be regarded as highly carbon-enriched (a *quasi*-martensite), in contrast to the ferrite in pearlite. It is most interesting to note that for instance Bain,[20] Zener,[21] and Jaswon[22] regarded the ferrite portion of bainite (in carbon steels) as supersaturated ferrite of equal carbon content to the parent austenite, a conception which would agree well with the present results on high-speed steel. Some difficulties were however pointed out by Mehl and Dubé.[23] That coherency can occur between austenite and the product bainite has been shown by Ko and Cottrell.[24]

The essential point, in the present context, is the suggestion that bainite formation might be considered as a projection at low temperatures of a process which fundamentally already occurs at high temperatures, namely in the region of the Ae_1 transformation, though on a different time-scale; the term ' high-temperature bainite ' might not be amiss, following upon the above ' high-temperature martensite '.

The so-called ' bainitic hardening ' process for high-speed steel due to Baer and Payson[25] (which produces a much improved cutting performance and consists of a heat-treating cycle of intermediate isothermal arrests at 235–315° C to form bainite and at 565° C to condition the retained austenite so that its further decomposition to bainite can proceed during the next holding at 235–315° C) may here be interpreted on the atomic scale in that it provides a means of controlled lattice-shear within the austenite crystal; it allows all γ-domains to transform fully without allowing precipitation to advance too far. That is, the resistance of austenite to transform is lowered but the isocompositional nature of the " ferrite " substantially maintained. In fact, in so far as the removal of residual γ-islands is aided by the movement of dislocations, it would seem possible to the author that the condition of oscillation in temperature would contribute to this end.

SUMMARY AND CONCLUSIONS

(1) The isothermal phase-changes during tempering of a quenched high-speed steel are followed by the high-temperature X-ray method at three tempering temperatures (600, 550, 500° C), and the austenite → martensite → ferrite decomposition observed with time.

(2) The decreased rate of decomposition with decreasing temperature is shown by the prolonged retention of austenite, and the retarded carbide formation and martensite/ferrite transition. At 500° C the progress is very slow but still significant (see Table I), the rate of austenite breakdown and carbide precipitation being considerably faster than the rate of martensite breakdown; the ratio of these two rates is the higher with decreasing tempering temperatures.

(3) The known fact that 550° C is the optimum practical tempering temperature for secondary hardening would accord with the conditions of disappearance of γ and of maximum martensite-formation, as here observed; not until this occurs does the major martensite → ferrite change set in.

(4) Martensite is found certainly to form isothermally (which however does not preclude its athermal formation in addition).

(5) The meaning of lattice expansion coefficients of the non-equilibrium (temper) phases is briefly considered; that of retained austenite should, from the results of Part I, be larger than that of ferrite, in agreement with bulk dilatometric knowledge; but the coefficient for martensite is a fundamentally indeterminate quantity.

(6) There is strong evidence for a continuous change in composition occurring during the $\gamma \longleftrightarrow \alpha$ transformation which suggests the concept of a saturated ' high-temperature martensite ' existing in the region of and just above the Ae_1 temperature, isocompositional with austenite, with an attendant ' high-temperature bainite.' Considering the phase diagram for high-speed steel in this light, its pro-eutectoid $\alpha + \gamma + \eta$ field would thus become in principle somewhat indeterminate, because the α/γ lattice-shear transformation is so labile that fluctuations could be tolerated amongst adjoining grains and even within a single grain between lattice domains (depending on local strain, lattice faults, and movement of dislocation). Phase hystereses are the ready consequence. Tempering reactions are viewed as simulations of the same high-temperature occurrences by the undercooled phases.

(7) The conception of the ferrite portion of bainite as a supersaturated phase (e.g. Zener) would accord well with the above.

(8) The limited composition variation of the matrix in 18/4/1 steel, and the relatively small atomic volume changes during the transformation favour such coherent α/γ coexistence. The high-speed steel martensite is regarded as a hybrid between the (Fe–Ni) and the (Fe–C) modes of martensite.

(9) Ferrite in nominally ferritic high-speed steel is able to retain traces of non-equilibrium solute, considered as a residue from its martensitic history. Such " pseudo-eqilibrium effects " are discussed; their presence or absence depends on pre-heat-treatment, and may well have a significant influence on practical high-speed steel behaviour.

Acknowledgment

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The Chlorine Method for the Determination of Non-metallic Inclusions in Steel

By C. W. Short, R. S. Roberts, and G. Croall

SYNOPSIS

This account of the determination of non-metallic inclusions in steel after separation by chlorine treatment describes two methods both of which give results in good agreement with vacuum fusion determinations on a wide variety of steels. 1306

THE CHLORINE METHOD for the determination of non-metallic inclusions in steel has already been very ably described by Colbeck, Craven, and Murray, whose original paper was introduced in 1936.[1] The present contribution offers some modifications to the procedure, a new apparatus for chlorinating the steel is described, and two separate schemes of analysis are suggested.

METHOD 1

Apparatus

The apparatus, corresponding very closely to that used by Colbeck, Craven, and Murray, consists essentially of a gas train, chlorination furnace, and ferric-chloride collecting vessels. The gas train is fitted up so that the apparatus may be flushed out with dry nitrogen or a measured flow of dry chlorine. Specially purified 'white spot' nitrogen, oxygen content less than 0·05 p.p.m., is used. The majority of any gaseous impurities in chlorine collect at the top of the cylinder and according to the manufacturers may be rejected by passing about 10% of the contents of a new cylinder into a sodium hydroxide absorber.

The chlorination assembly is illustrated in Fig. 1. All standard glass joints are fitted with hooks and springs and are well greased with Vaseline. It is important that the space between the outside wall of the reaction tube and the inner wall of the B.55 joint be as large as possible. If any constriction of the glass takes place at this point the space may become choked with ferric chloride and prevent free passage of chlorine.

The Nichrome element of the heating furnace is wound so that the furnace has a hot zone at one end. The insulating bricks covering the winding are loose and can be removed as also can be the top and one side panel. Rapid cooling of the furnace tube can then be effected by means of a small electric fan.

Preparation of Sample

Steels for chlorination are usually received in bar, ingot pieces, tube, or strip form. The size of the sample must be such that the chlorination can be

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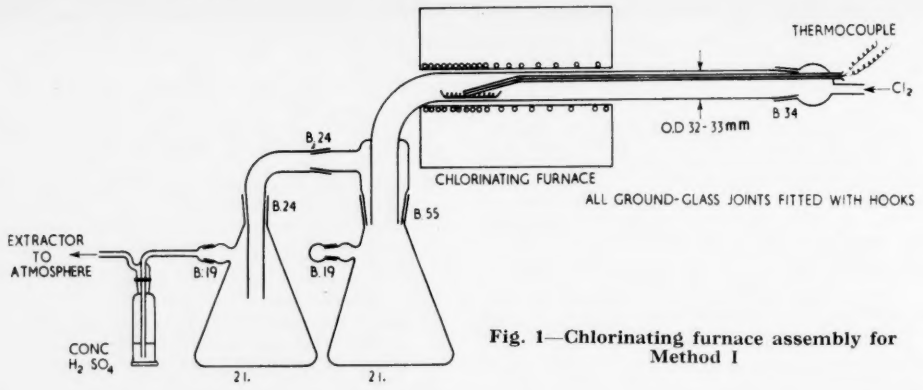


Fig. 1—Chlorinating furnace assembly for Method I

completed in a maximum of 3 to 4 h, thus enabling the operator to prepare both the apparatus and the sample, to chlorinate, filter the residue, and to wash up the glass vessels in one working day. Generally speaking, a useful weight of sample is 10 to 13 g. This weight will give a reasonable amount of residue for subsequent analysis, and will chlorinate in about 3 h, providing that the following sampling technique is followed.

Wherever possible, samples should be rolled or forged into a $\frac{3}{8}$-in. section, then lightly skimmed to remove adhering scale. Discs $\frac{1}{10}$ in. thick should be parted off, 9 such discs usually being sufficient for one chlorination.

If it is not practicable to roll or forge down to $\frac{3}{8}$-in. section, $\frac{1}{10}$-in. slices across the face of the sample should be taken and cut up, preferably by guillotine, to fit the boat, care being taken that the total sample weight represents so many complete cross-sections. Samples in tube form are similarly skimmed inside and outside, a $\frac{3}{8}$-in. ring cut off, and this ring cut up into small sections about $\frac{3}{8}$ in. long $\times$ $\frac{1}{4}$ in. wide $\times$ $\frac{1}{10}$ in. thick. Samples in strip form are lightly machined, filed, or buffed, or in the case of high silicon strip, pickled in hot 50% hydrofluoric acid, washed in hot dilute ammonia, then hot water and rinsed in alcohol.

It is important that there be an absolute minimum delay between preparing the sample and introducing it into the chlorinating apparatus. If some delay cannot be avoided, the prepared sample should be stored in acetone, ether, or absolute alcohol.

Chlorination Procedure

The normal procedure for the actual chlorination is followed, but the rate of flow of chlorine is regulated at 20 l./h. If an excessive chlorine velocity is used then ferric chloride is apt to be carried beyond the settling vessels. When the furnace temperature has reached 250° C the power input is carefully controlled until the temperature remains steady at 400° C. Chlorination is complete in about 3 h, when the furnace is switched off. At room temperature the apparatus is swept out with pure dry nitrogen and the sample boat removed for subsequent treatment of the residue.

Treatment of Residue

The flocculent residue is extracted with a little cold water, filtered under slight suction through a paper pulp pad and washed in the following sequence:

1. 100 ml. boiling water
2. ,, hot 3% sodium carbonate
3. ,, boiling water
4. ,, cold water
5. ,, cold 5% hydrochloric acid
6. ,, cold water

The sodium carbonate wash is used to remove hydrolysis products of any silicon compounds and the hydrochloric acid to remove ferric chloride and sodium compounds. After washing, the residue is ignited at 1000° C to constant weight.

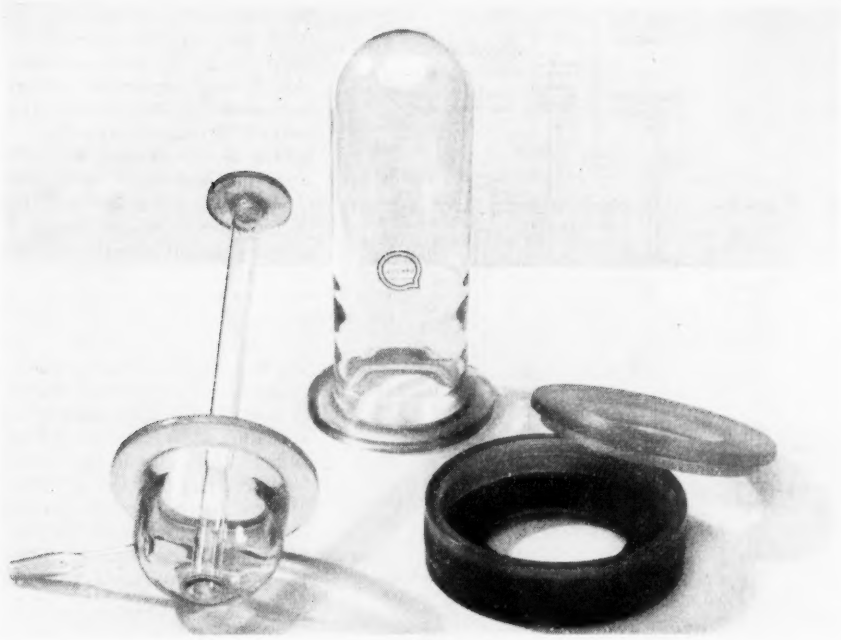
Scheme of Analysis

The following is the general scheme of analysis of the ignited residue:

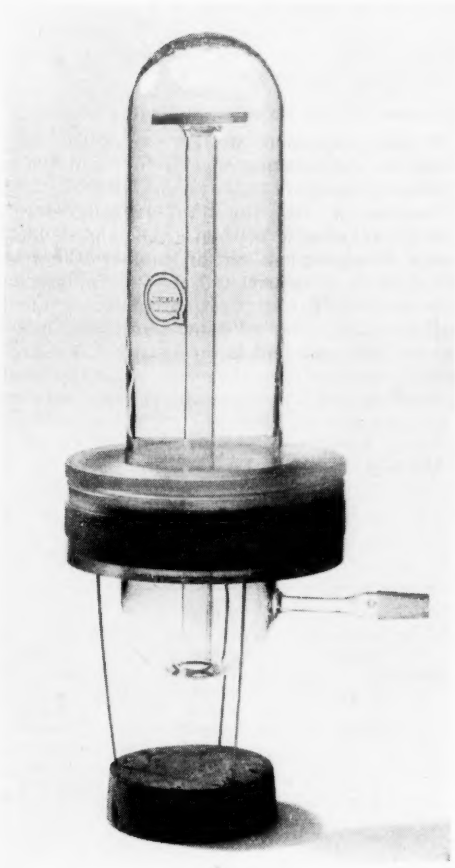
(i) Solution of residue in hydrochloric acid and perchloric acid or nitric acid and perchloric acid*
(ii) Separation and determination of silica
(iii) Precipitation of Fe, Ti, Zr, by means of cupferron
  (*a*) Iron determined colorimetrically using thioglycollic acid
  (*b*) Titanium determined calorimetrically using H_2O_2
  (*c*) Zirconium determined gravimetrically as phosphate
(iv) Precipitation of Al and Cr as hydroxides, after reduction with sulphur dioxide. Aluminium determined using stabilized aluminon reagent. Chromium determined on separate aliquot using diphenyl carbazide reagent.
(v) Precipitation of manganese as hydroxide, following oxidation with bromine. Determined colorimetrically using potassium periodate.
(vi) Precipitation of Ni by dimethyl glyoxime. Determined volumetrically.
(vii) Phosphorus pentoxide is very rarely present in quantities over 0·001 % and is not normally determined. If necessary, it is determined colorimetrically using stannous chloride and ammonium molybdate reagents.

However, if an accurate chromium content of the residue is required in order to check on the completeness of the analysis then the initial solution of the residue can be made by nitric acid and perchloric acid.

* Whilst the residue is being dissolved and oxidized by perchloric acid there is a possibility of loss of chromium due to volatilization of chromyl chloride. Now the Cr present in the chlorine residue is not considered as an oxide-bearing element in the final oxygen results so that loss of Cr at this stage does not affect the total oxygen determination.



(*a*)



(*b*)

Fig. 2 — (*a*) and (*b*): apparatus for Method II

METHOD II

Apparatus

The apparatus, illustrated in Figs. 2*a* and *b*, is made entirely of Pyrex glass and consists of two main parts. The bottom component is a $3\frac{1}{2}$-in. outside dia. tube, one end fitted with a $\frac{3}{4}$-in. flat flange, the other end being closed and fitted with a $\frac{5}{8}$-in. bore re-entrant tube which terminates in a flat sample platform 2 in. dia. The tube wall has a side-arm dip-tube, the external end terminating in a B 14 cone. The top component consists of a $3\frac{1}{2}$-in. tube, with a $\frac{3}{4}$-in. flat flange to match the bottom component, and the other end closed hemispherically. The overall height of the apparatus is about 15 in.

The two flanges are well greased with vacuum grease before assembly and are then clamped together by means of a union-type ring, rubber gaskets being used between the glass and metal faces. This ring can be made satisfactorily in Perspex.

Chlorine is obtained from a 10-lb capacity cylinder through a silver pipe via a pressure gauge and reducing valve. The gas is dried by passage through a drying train consisting of sulphuric acid, calcium chloride, and phosphorus pentoxide towers. Between the drying train and reducing valve, a **T**-piece is inserted, connected to a rubber bladder which acts as a reservoir. When the chlorine leaves the drying train it can be passed into either the reaction chamber or a caustic-soda absorber, the latter path being used to clear all air from the drying train before allowing chlorine to enter the reaction chamber. Figure 3 shows a block diagram of the complete unit and is self-explanatory. The vacuum pump must be capable of reducing the pressure to about 0·01 mm Hg, and a single stage rotary oil pump, ultimate vacuum 0·005 mm Hg, is quite suitable. The vacustat is used to measure the pressure in the reaction chamber and a suitable instrument covers the range 1·0–0·001 mm Hg.

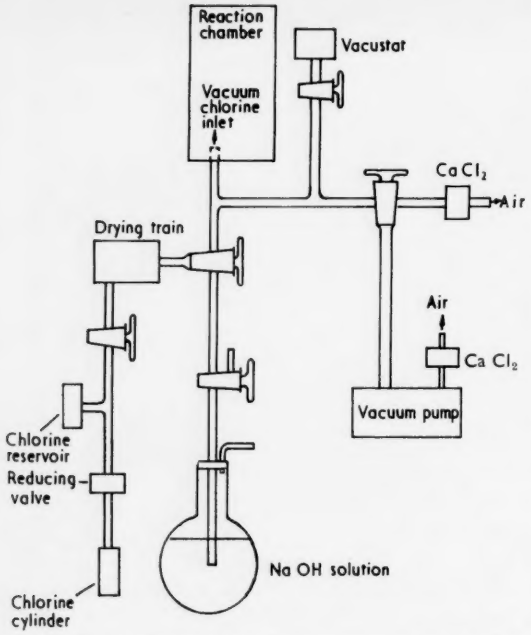


Fig. 3—Block diagram showing connections for Method II

Preparation of Sample

The samples are prepared as in Method I and used in the form of cylinders $\frac{7}{8}$ in. long by $\frac{3}{8}$ in. dia., not in the form of thin discs.

Procedure

After the previous chlorination the apparatus will have been thoroughly cleaned and baked in an oven at 150–200° C. The sample carrier consists of a silica capsule 2 in. dia. and $\frac{1}{2}$ in. deep. It is fitted with a lip to facilitate subsequent removal of the chlorinated residue by washing. The clean, dry, weighed sample is placed in the capsule which is then placed on the sample platform and the apparatus assembled. The system is evacuated by means of a rotary oil pump, the degree of vacuum being obtained by readings on the vacustat. When the pressure is reduced to 0·01 mm and the apparatus confirmed as vacuum tight the pump and vacustat are isolated from the chlorination apparatus. Meanwhile, chlorine is led from the cylinder through the drying train into the sodium hydroxide absorber. When the connecting tubing is well purged with chlorine and the bladder reservoir well-filled with chlorine, the cylinder is closed. Then the two-way tap controlling the flow of chlorine from the reservoir is slowly turned to admit chlorine into the reaction chamber. The temperature is slowly raised and evolution of chlorides commences between 200° and 250° C; the temperature is finally raised to 350° C and carefully controlled at this figure by means of a Sunvic energy regulator. The chlorine reservoir is filled as necessary, from the cylinder. The time of chlorination is about 6–7 h: in practice the chlorination is started during the day and allowed to continue overnight. There is normally sufficient chlorine in the bladder to allow this.

When chlorination is complete the apparatus is allowed to cool and is dismantled, the sample being immediately removed for the next operation. The chlorination apparatus is then cleaned and baked ready for future use.

Treatment of Residue

Upon completion of chlorination, the residue is treated as described in Method I. After the final ignition, however, it is analysed by absorptiometric methods which have been described in detail by Greenfield and Sparrow.[2]

Applications of the Chlorine Method

Extensive experience in the use of the chlorine method has shown that, on a wide variety of plain and alloy steels, results may be obtained which are in close agreement with the total oxygen determined by the vacuum fusion method. The chlorine method, however, is subject to a certain number of limitations which do tend to restrict its scope of application.

It is essential that all steels for chlorination be fully killed; rimming steels give low results because some of the inclusions present in these steels are susceptible to attack by chlorine at the chlorination temperature. This has long been a problem with the chlorine method,

Table I

COMPARISON OF CHLORINE AND VACUUM FUSION METHODS

Steel Type						Total Oxygen, %	
Al, %	C, %	Si, %	Ni, %	Cr, %	Mo, %	Vacuum Fusion Method	Chlorine Method
	0·23	0·15				0·010	0·011
	0·065	0·04	3·01	0·66		0·026/0·029	0·025/0·029
	0·21	0·14	0·60	0·12	0·26	0·020	0·021
	0·01	(Heavily treated with aluminium)				0·003/0·004	0·0035/0·005
	0·31	0·17	3·76	0·91	0·24	0·004	0·005
	0·32	0·17	3·06	0·84		0·004	0·004
	0·30	0·30	2·37	0·84	0·45	0·004	0·006
0·15	0·28	0·18	0·085	0·015		0·0095	0·0105
0·67	0·28	0·19	0·085	0·005		0·011	0·013
	0·30	0·25	0·18	2·98	0·63	0·002	0·004
	0·18	0·15	0·41	5·38	0·55	0·0045	0·007
	0·34	0·12	0·42	12·96		0·009	0·011
0·50		4·00				0·002	0·004

Table II

ANALYSIS OF CHLORINE RESIDUES

No.	Sample	Chlorine Method	Temp., °C	Residue Analysis with each constituent expressed as % of steel sample					Total Oxygen, %	Total O_2 by Vacuum Fusion
				SiO_2	FeO	MnO	Al_2O_3	Cr_2O_3		
1	Basic Bessemer Rimming Steel Cast 8652 0·05% C; 0·44% Mn	I	350	0·0015	0·010	0·005	0·004		0·006	0·0145
		I	400	0·001	0·009	0·001	0·002		0·004	
		II	350	0·001	0·001	0·001	0·002		0·002	
2	Heavily Al-killed Basic Bessemer Cast 4654 0·01% C, other elements low	I	400	0·003	0·004	trace	0·002		0·003	0·004
		I	400	0·004	0·008	trace	0·002		0·005	
		II	350	nil	nil	trace	0·0074		0·0035	
		II	350	nil	nil	nil	0·008		0·0035	
3	Basic O.H. Cast 784 0·23% C; 0·66% Mn; 0·15% Si	I	350	0·006	0·022	0·004	0·004		0·010	0·010
		I	400	0·007	0·014	0·007	0·004		0·010	
		II	350	0·012	0·001	0·004	0·006		0·010	
		II	400	0·010	0·002	0·002	0·007		0·009	
4	B.O.H. Cast 7678 0·20% C; 0·65% Mn	I	350	0·003	0·009	nil	0·010	0·012	0·007	0·006
		I	400	0·002	0·007	trace	0·011	0·002	0·0065	
		II	350	0·002	0·001	trace	0·013	0·009	0·007	
		II	400	0·002	0·001	trace	0·013	0·001	0·007	
5	E.F. Steel No. 1269 0·30% C; 0·62% Mn; 0·17% Si; 0·60% Cr	I	350	0·009	0·010	trace	0·015	0·069	0·012	0·009
		I	400	0·006	0·009	trace	0·014	0·007	0·010	
		II	350	0·006	0·001	trace	0·014	0·051	0·010	
		II	400	0·007	nil	trace	0·014	0·008	0·010	

and, as yet, no solution to it has been found.

If the carbon content of the steel is greater than 0·5%, erratic results are obtained, due mainly to the difficulty in washing chlorine residues containing large quantities of carbon.

Chromium steels, chlorinated at the original temperature of 350° C, often proved quite troublesome. Large quantities of chromium carbides were included in the oxide residues and created some difficulties in the scheme of analysis; fuming with sulphuric acid caused precipitation of the comparatively insoluble chromic sulphate. Chlorination at the higher temperature of 400° C largely overcomes the difficulty. Most of the chromium is converted to chloride and volatilized along with the other volatile chlorides. Any chromium left behind in the residue is oxidized by perchloric acid to dichromate, thus simplifying the analytical procedure. It must be remembered that only the oxides of Si, Fe, Mn, and Al are used for the final oxygen count; thus for accurate analysis of a few milligrams of residue these oxides should constitute the major proportion of the residue. Furthermore, as only killed steels are chlorinated we may expect the predominant elements in the residue to be silicon and aluminium and since the oxides of these two elements contain about 50% oxygen, a high degree of accuracy is essential in the analysis of the residue.

Some typical results obtained by the chlorine method are shown in Tables I and II.

Table I illustrates the comparative results for total oxygen obtained by the vacuum fusion and chlorine methods. It will be seen that good agreement is obtained on a representative number of carbon and alloy steels. Generally speaking, the total oxygen obtained by the chlorine method is 0·001–0·002% higher than the vacuum fusion oxygen.

Table II illustrates some detailed results of residue analysis and gives a brief indication of the applicability of the method.

Sample No. 1 is a basic Bessemer rimming steel and emphasizes the major shortcoming of the chlorine method, average total oxygen being only 0·004% against the vacuum fusion figure of 0·0145%.

The remainder of the specimen results show good agreement with the total oxygen determined by vacuum fusion.

Samples 4 and 5 illustrate how an increase in the chlorination temperature from 350° to 400° C will successfully reduce the chromic oxide content of the residue.

A significant feature of the samples 2–5 is the difference in iron oxide content when determinations are made by the two chlorination methods. It may be recalled that the vacuum apparatus of Method II will use considerably less chlorine than the conventional apparatus of Method I. It is therefore a possibility that a minute concentration of oxygen remains in the chlorine cylinder after the initial bleeding off of impurities and that this oxygen combines with iron and produces iron oxide, which is included in the residue for analysis. However, this is a matter of some conjecture and one well worthy of future study. Small quantities of oxygen in chlorine are not easily or conveniently removed and this could possibly be an important factor in favour of the apparatus described in Method II.

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The Cooling of Hot Steel Strip with Water Jets

By A. Sigalla

SYNOPSIS

Two series of experiments have been undertaken on the cooling of hot steel strip and their relevance to the design of strip cooling systems is discussed. It has been found that rapid cooling can be achieved using simple water jets and a formula is given from which the approximate number required can be calculated. 1392

Introduction

VARIOUS ARRANGEMENTS and types of water sprays are employed in steelworks for the cooling of hot steel strip on the runout table, but the control that can be maintained when strips of different thickness and temperature are delivered to the cooling zone is not always adequate. Furthermore, the quantities of water employed appear to be disproportionate to the amount of heat that it is required to remove from the strip. For these reasons a study of the heat transfer from a moving steel strip has been undertaken with the object of providing a rational basis for design of strip cooling systems that would permit closer control of strip temperature at the coiler and more economical use of water.

PRELIMINARY EXPERIMENTAL WORK

Owing to the large number of variables involved in the study of the heat transfer from a moving strip, and the considerable number of experimental difficulties inherent in the reproduction of runout conditions in the laboratory, the investigations began with a preliminary series of qualitative experiments.[1] These were designed to illustrate the relative importance of the variables affecting this type of heat transfer, and together with a study of rolling mill operational data enabled the construction of more accurate experimental apparatus.

Specimens of mild-steel plate, about 6 in. × 6 in. and $\frac{1}{2}$ in. thick having thermocouples embedded from the underside with the junctions very close to the upper surface, were heated to a uniform temperature of about 1000° C, and the coolant was directed at the centre point of their upper faces.

The cooling systems studied were (*a*) coarse and fine water sprays, and (*b*) water jets.

It was observed that water jets appeared to cool the plate more rapidly than either the coarse or fine water sprays.

Because of the slow response of the temperature recording apparatus, the numerical results were mostly of a qualitative nature, but these were sufficient to establish a basis of comparison.

The water droplets from the sprays seemed incapable of effectively contacting the plate, for the finer droplets were sometimes diverted from it by air convection currents, and those droplets arriving at the surface immediately coalesced into sheets which ran off the steel plate. By contrast, the effect of water-jet cooling was characterized by the formation of a ' black spot ' on the hot steel surface at the point of jet impact. This black spot appeared almost instantaneously and thereafter increased in size until the whole plate was cooled. Provided that the water flow rate was sufficient to maintain the jet formation, an increase of water flow rate did not materially increase the rate of heat removal.

These observations may be discussed bearing in mind the steam layer which insulates hot surfaces from further cooling when the temperature of the cooled surface appreciably exceeds the boiling point of water. This state of affairs is then quite different from nuclear boiling and indeed the heat fluxes are very much reduced. Typical cooling curves are given elsewhere,[2, 3] as is a brief yet very lucid description of the phenomenon.[4]

The ordinary water spray is probably an effective coolant only during its initial moment of application. Then the droplets coalesce into the water sheets insulated from the hot surface by the steam layer. This behaviour can often be seen on strip rolling mills. In general, the use of higher water pressures will not cause the spray droplets to penetrate this sheet of water since although their nozzle velocity may be higher, the droplets are smaller and their velocity retardation in travelling to the hot steel surface may be greater. A jet maintains a continuous supply of the coolant at the surface and appears to prevent the formation of a steam layer at its point of impact.

Apart from the small zone at the point of impact, the flow from the jet can be considered approximately as plane tangenital,[5, 6] and a theoretical study[7] of the heat transfer under such conditions reveals that a surface temperature distribution is formed which will favour the formation of a steam layer. It is

Paper MW/A/34/56 of the Rolling Committee of the Mechanical Working Division of the British Iron and Steel Research Association, received on 29th January, 1957. The views expressed are the author's, and are not necessarily endorsed by the committee as a body.

Mr. Sigalla is with the Association's Physics Department at Battersea, London.

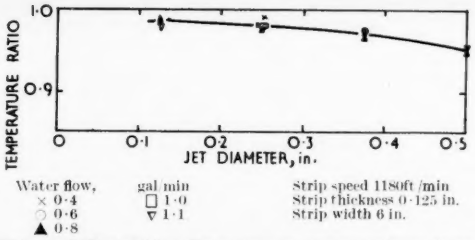

Fig. 1—Effect of jet diameter on temperature drop

also shown that large changes in the external water flow will have very small effects on the temperature distribution. All this is in broad agreement with the experimental observations, which were further confirmed by an analysis of rolling mill data.[8] From this data it was possible to obtain an empirical first order relationship between the more important variables without taking into account changes in water volume per unit area of strip.

MAIN EXPERIMENTAL WORK

The experiments were undertaken to verify that the observations made and conclusions drawn from the preliminary static experiments were valid when applied to a moving strip of steel, and obtain quantitative information, relating the cooling effect to the relevant variables.

The apparatus consisted of a 6-in. annulus of steel strip, 4 ft 3 in. outside dia., rotated at various speeds, and passing through a tunnel furnace and a cooling zone. The furnace covered half the circumference of the annular strip and the cooling zone the other. Heat was supplied by three 'Base-Quip' burners. The annulus was driven by a 4-h.p. electric motor, and a gearbox and pulley system made possible changes of angular speed from 54 to 157 rev/min.

Strip surface temperatures were recorded by means of a calibrated Brown Radiamatic radiation pyrometer with a $1\frac{3}{8}$-in. dia. target, mounted vertically above the annulus, in conjunction with a single-point temperature recorder and a chart speed of $\frac{1}{3}$ in./min.

Arrangements were also provided for the testing of a number of nozzles and sprays in a pressure range up to 90 lb/in^2 and water flows up to 2 gal/min per nozzle.

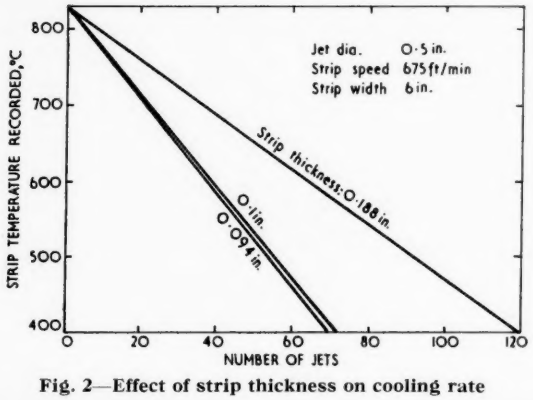

Fig. 2—Effect of strip thickness on cooling rate

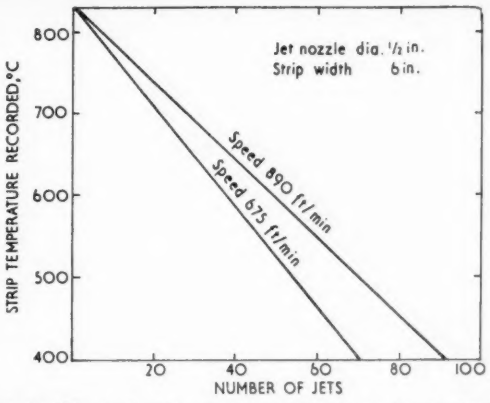

Fig. 3—Effect of strip speed on cooling rate (strip thickness 0·94 in.)

Experiments were conducted in two ways:

(i) Temperature measurements were taken at fixed positions upstream and downstream of the cooling section after a steady state had been established. The aim of these experiments was to determine the effect of the water jet diameter and flow rate on the temperature drop for fixed strip thickness and speed.

(ii) With no heat input and the strip being steadily cooled by the water jet, a temperature time record was taken to determine the cooling effect over a range of strip thicknesses, speeds, and water jet diameters.

The results of (i) above are shown in Fig. 1. The independent variable, (which can be interpreted as the cooling effect), is the ratio of the downstream temperature to the upstream temperature. This method of representation was found preferable to a plot of temperature differences as otherwise it would have been necessary to deal with differences of quantities of the same order of magnitude. The mean temperature of the strip in these experiments was of order 800° C. Figure 1 shows the temperature ratio plotted against cooling jet diameter for a range of flow rate. It is seen that cooling improves with increasing jet diameter and that most points tend to fall on one curve (strip speed and thickness were the same for all experiments). Though the measured changes were relatively small, the trend is well defined. Figure 1 also shows that any effects which might have been caused by changes in water flow were too small to be detected by these experiments. This was found to agree with the results of the preliminary experiments and the other studies. Figures 2 and 3 are representative of the effects of the strip thickness and speed on the cooling rate.

The ordinate is the temperature measured at a point well downstream of the cooling zone. The abscissa for both curves is the number of times a strip element has passed under the water jet, calculated from the rotational speed of the strip annulus. These curves were copied from the recorder charts.

They show that the cooling effect is decreased by increasing the strip thickness and speed, which is to be expected.

The measured temperature changes included not only the heat removed by the water jet but also the extraneous losses. The heat losses were of appreciable

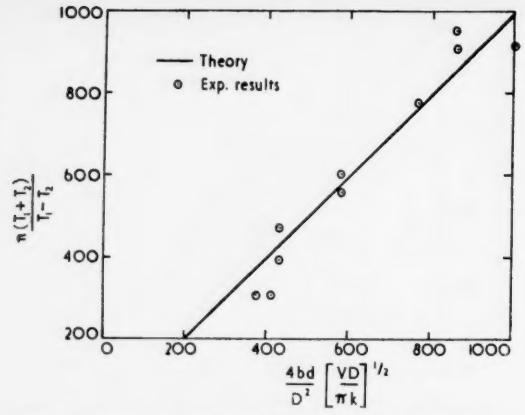

Fig. 4—Non-dimensional representation of results

magnitude and rather difficult to estimate. Two tests were therefore carried out for each trial; one with the strip undergoing water cooling and another 'blank' test, with no water cooling. For given conditions it thus became possible to obtain by difference the net cooling effect of one water jet. Figure 4 is a non-dimensional plot of all the test results, corrected for the extraneous heat losses which were determined in the manner described.

The ordinate is:

$$\frac{n(T_1 - T_2)}{T_1 - T_2}$$

where T_1 = temperature at the start of the cooling period, T_2 = temperature at the end of the cooling period, n = number of times a strip element has passed under the jet during cooling time.

This grouping of variables was chosen because calibration errors in the temperature measurements are thus minimized and the representation of the experimental data in this way resulted in a linear plot when plotted against the non-dimensional abscissa:

$$\frac{4bd}{D^2}\left[\frac{VD}{\pi K}\right]^{\frac{1}{2}}$$

where b = width of strip portion cooled by one jet, d = thickness of strip, D = diameter of water jet at or near point of impact, V = linear strip speed, K = thermal diffusivity of metal.*

Figure 4 can be interpreted to give the following relation for the number of jets, in line with direction of travel, required to cool a strip through a given temperature drop for given rolling conditions:

$$n = \frac{4}{\pi^{\frac{1}{2}}}\,\frac{bd}{D^2}\left(\frac{T_1 - T_2}{T_1 + T_2}\right)\sqrt{\frac{VD}{K}}$$

THEORY

It is possible to give an approximate theoretical explanation of the above if it is assumed that an impinging water jet maintains the surface temperature of the strip to that of the water over an area equal to the jet cross-section at the place of impact. This assumption, which was suggested by the experiments, permits calculation of the number of jets arranged in the line of motion of the strip which are required to cause a given temperature reduction for a known strip and rolling conditions. Direct experimental verification of this assumption was not found possible, but its justification lies in a comparison of deduced consequences with experiments.

The travel speeds of strip on a runout table are of order 1000 ft/min and the diameter of the jet at the point of impact is of the order of 0·5 in., therefore the time interval during which an element of strip is cooled by one jet is of order $2{\cdot}5 \times 10^{-3}$ s. The depth to which a temperature disturbance can diffuse in that time is less than the strip thickness and the strip is therefore initially considered as a semi-infinite body.

The heat transfer from a semi-infinite body, initially at a temperature T and instantaneously cooled at its surface to θ, is equal[9] to q where:

$$q = k(T - \theta)\sqrt{\pi K t} \qquad \ldots\ldots(1)$$

In this equation, k = thermal conductivity of body, K = thermal diffusivity of body, and t = time.

If the diameter of the jet is D, then the heat Q removed from an area $\pi D^2/4$ during a time interval τ is given by

$$Q = \pi D^2 \tfrac{1}{4}\int_0^\tau q\,dt = \frac{\pi D^2}{2}\,k\left(T - \theta\right)\sqrt{\frac{\tau}{\pi K}} \qquad \ldots\ldots(2)$$

Now consider an element of strip of width b, thickness d, and length D travelling at a speed V. The time spent by every such element under one jet is:

$$\tau = D/V \qquad \ldots\ldots(3)$$

Therefore the heat removed from the element by one jet is given by:

$$Q = \frac{\pi D^2}{2}\,k\left(T - \theta\right)\sqrt{\frac{D}{\pi K V}} \qquad \ldots\ldots(2a)$$

Let the initial and final temperatures of this element of strip be T_1 and T_2. The heat that it is required to remove from the element is then H, where:

$$H = (\rho c)(b.D.d)(T_1 - T_2) \qquad \ldots\ldots(4)$$

where ρ = density of element, c = specific heat. From equation (2a) the heat removed by one jet is approximately given by:

$$Q = \frac{\pi D^2}{2}\,k\left(\frac{T_1 + T_2}{2} - \theta\right)\sqrt{\frac{D}{\pi K V}} \qquad \ldots\ldots(5)$$

where the temperature T in equation (2a) has been replaced by $(T_1 + T_2)/2$, the mean temperature of the strip as the strip travels along the runout. This implies a linear temperature drop, which in fact was recorded during the experiments. Also, to the degree of accuracy of this investigation, we are justified in ignoring the contribution of θ to equation (5) which then becomes:

$$Q = \frac{\pi D^2}{2}\,k\left(\frac{T_1 + T_2}{2}\right)\sqrt{\frac{D}{\pi K V}} \qquad \ldots\ldots(5a)$$

* The value of K employed in this work was 0·075 cm²/s. This is an average value for steel in the relevant temperature range.

Now if n jets are required to cause the necessary temperature drop,

$$nQ = H \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (6)$$

And on substitution and re-arrangement we obtain

$$n = \left(\frac{4bd}{D^2\pi^{\frac{1}{2}}}\right)\left(\frac{T_1 - T_2}{T_1 + T_2}\right)\left(\frac{VD}{K}\right)^{\frac{1}{2}} \quad \ldots\ldots\ldots\ldots (7)$$

This equation was found to explain the experimental data.

CONCLUSIONS AND COMMENTS

It was found that rapid cooling of moving strip can be achieved by the use of simple water jets. Experiments and theory have also led to a formula which can be used to compute the number of jets to cause a required temperature drop for given rolling conditions.

The number of jets n in the formula

$$n = \frac{4bD}{\pi^{\frac{1}{2}}D^2}\left(\frac{T_1 - T_2}{T_1 + T_2}\right)\sqrt{\frac{VD}{K}}$$

represents only the number of jets in the direction of motion of the strip. The lateral number of jets is implicit in the factor b on the right hand side, which is the lateral length allotted to a single jet.

Nevertheless, it is the number of jets in the direction of motion which is the more important quantity since this is the parameter which decides the actual length of runout to be made available for cooling purposes.

Another simple conclusion that can be drawn from the above result, is the desirability of staggering the jets in the direction of motion, this in effect reduces b and n. Experiments[10] have also shown that it is possible to reduce the value of b, without undue jet interference effects.

Metallurgical analysis of low-carbon steel segments subjected to cooling of the type discussed in this report has shown that it had no detrimental effect on the quality and structure of the steel.

APPENDIX

Application of Results to Continuous Casting

A derivation suggested by Savage[11] has made it possible to extend the formula

$$n = \frac{4bD}{D^2}\left(\frac{T_1 - T_2}{T_1 + T_2}\right)\sqrt{\frac{VD}{K\pi}}$$

to cases of continuous casting. A new formula is thus deduced giving the rate of heat removed by one jet under such conditions.

Starting with the previously derived formula for the amount of heat Q removed by one jet

$$Q = \frac{\pi D^2}{2}k\left(\frac{T_1 - T_2}{2}\right)\sqrt{\frac{D}{\pi KV}}$$

For one jet, $T_1 \simeq T_2$, so that $(T_1 - T_2)/2 \simeq T$ temperature of billet.

Then,

$$Q = \frac{kT}{2}\sqrt{\frac{\pi}{K}\frac{D^5}{V}}$$

And rate of heat removal, i.e. heat removed per unit time, is given by

$$\frac{QV}{D} = H_f, \text{ and } H_f = \frac{kT}{2}\sqrt{\frac{V\pi D^3}{K}}$$

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DISCUSSION ON PAPERS

Discussion at Meetings · Written Contributions

CORROSION MEETING, 1956

By arrangement with The British Iron and Steel Research Association, a Corrosion Meeting organized by The Iron and Steel Institute was held at the offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, on Friday, 12th October, 1956. The Chair was taken by Dr. H. H. BURTON, C.B.E., President of the Institute, supported by Dr. J. PEARSON, Assistant Director of the Association, who occupied the Chair during part of the afternoon session.

Papers were presented at both sessions and the discussions are given below.

CORROSION RESISTANCE OF LOW-ALLOY STEELS

This discussion was based on the paper "The Corrosion Resistance of Low-Alloy Steels" *by J. C. Hudson and J. F. Stanners* (J. Iron Steel Inst., 1955, vol. 180, pp. 271–284).

Dr. **J. C. Hudson** (B.I.S.R.A.) presented the paper.

Mr. **T. Henry Turner** (British Transport Commission): There is no internationally accepted definition of a low-alloy steel. When I suggested these tests I had in mind slow-rusting steels which included deliberate additions of alloying elements, to enhance their mechanical properties for many general engineering purposes for which wrought iron or mild steel would have been used in the recent past.

Among the 60 steels which the Committee had made for their tests I am now surprised to find how relatively few 1·5% Mn steels appear. 3% of Ni or Cr would be unobtainable or too expensive for the type of steel I had in mind. Steel No. 20 is certainly a possibility and now widely made in the U.K. The rate of corrosion given by the authors for steel No. 20 is only 2·0 whereas the reference steel is about 6·0 which shows a really useful improvement in corrosion resistance for this type of low-alloy steel. These corrosion tests might have been completed and reported ten years ago but for the war. They were planned and financed through the Institute's Corrosion Committee under the leadership of Dr. W. H. Hatfield whom the older members of that Committee recall with respect and affection. That Committee, now under B.I.S.R.A., held its 135th meeting yesterday.

The tests were underway before the formation of B.I.S.R.A. and when there was no Corrosion Group of the Society of Chemical Industry. It is remarkable that the whole series of tests was not abandoned in the stress of war.

When Dr. Hudson found difficulty at any of the Committee members' works in arranging for the manufacture of these special low-alloy steels, or their rolling, or the machining of specimens etc., he always managed to find some other member of the Committee whose firm's staff took over part of the work. Dr. Hudson remained throughout all those years the essential centre for co-ordinating some very complicated scientific team work.

Now that this research has been completed and reported to us, I suggest that we should take the following action.

(i) Record, for the benefit of steel users, the steels found to resist corrosion better than more ordinary steels, in so far as such steels may be said to comply with the compositions tabulated in any B.S.I. list. Some B.S. specifications will, I hope, refer in the future to the slow-rusting of some grades of steel as a feature of value.

(ii) Try to apply to these steels the Atmospheric Corrodibility Index recently put forward by the S.C.I. Corrosion Group. If the corrodibility index is not yet finalized this would be a good try-out for it.

(iii) Record the changes in steelmaking practices which have occurred during the 20 years since these tests were initiated. It may well be that the scope of the low-alloy steels tested has been changed by the wide use of oxygen in steelmaking and the outstanding developments in steel sheet rolling, to mention only two aspects.

(iv) Investigate the changes which have occurred through copper and other residuals entering with the scrap used in making such steels. Whereas copper was a deliberate addition in some slow-rusting steels, it is now appearing in medium carbon steels, and by concentrating in the skin during the oxidation which occurs while forging and reheating it may become a nuisance penetrating the steel at high temperatures.

(v) Indicate for a given gauge of wire or sheet what the increase in cost would be per ton or per mile or per square yard, whichever be the most useful, for these slow-rusting steels as compared with present-day mild steel which often contains appreciable quantities of copper.

During these tests I learnt to admire the zeal and scientific honesty of both the authors.

I think I supplied the steels for the atmospheric corrosion test results recorded in Table III, p. 278.

because I knew that corrosion accelerated serious corrugation and that serious corrugation had been noted especially on rails with the running surface cooled during manufacture more quickly than usual to produce a harder surface with a sorbitic type of microstructure. According to my hypothesis, the corrosion tests should have shown that the sorbitic rail steels corroded more rapidly than ordinary rail steel, but in fact the tests showed little difference in the rates of corrosion of the four types of rail steel. Perhaps it is the better wearing quality of the sorbitized and dried 'hill' which causes it to remain for a longer time proud of the wet and corroding troughs on either side of it in the corrugating rail top surface.

Dr. **S. A. Main** (Hadfields Ltd.): May I add to Mr. Turner's remarks on the early investigations of corrosion of alloy steels in sea-water and in other media. In their programme of research which was instituted well before the setting up of the I.S.I. Committee, the Institution of Civil Engineers covered ferrous alloys with a number of alloying elements as well as wrought iron. The research was carried on by the Sea Action Committee of the Institution for 15 years with successful results.

Dr. **T. P. Hoar** (Department of Metallurgy, University of Cambridge): I am well aware of the difficulties of making the more expensive low-alloy steels, but it seems to me that they have been overcome to a large extent in countries other than the U.K., particularly the U.S.A. I find it difficult to see why they cannot be overcome here. I hope that this paper will lead to a reappraisal of the possibilities of the use and manufacture of these low-alloy steels.

What is the reason for the considerably increased corrosion resistance of this type of material as compared with that of mild steel? There seem to be at least two possibilities. The first is that the material is itself inherently less easily dissolved or converted into compounds. The second is that the compounds formed have considerably more protective action on the metal remaining below them. Perhaps the authors could say a little more on the mechanism by which the resistance of the low-alloy steels is achieved. The production of compact corrosion products has long been believed to be an important reason why certain steels corrode less rapidly than those that produce less compact corrosion products, but this may not have been generally appreciated by non-specialists in corrosion.

Dr. **F. Wormwell** (Chemical Research Laboratory, Teddington): It is most important that the best possible use should be made of the large amount of material that has now been obtained from these very well and very patiently conducted researches.

Mr. Stanners mentioned the degree of localized corrosion that may sometimes occur on these steels. In general, the results as given in the paper are in terms of average loss of thickness based on weight loss of the specimens, and I believe Mr. Stanners did say that in the immersion tests at least there was very little sign of localized pitting. Would he confirm whether that was actually the case, because in some work at Teddington there has sometimes been found a tendency to localized pitting on some of the steels, usually those containing about 3% Cr.

Some work was done at Teddington on some of these steels, but unfortunately we were unable to complete the original programme. However, in general our results confirm the conclusions drawn by the authors.

Some experiments were done in immersed conditions and corrosion/time curves[1] were obtained on a few of the steels in stagnant conditions. Corrosion against time was measured by oxygen absorption. For some of these steels, including No. 46, which contained about 3% Cr, in our experimental conditions in half-normal sodium chloride the steels did corrode at about the same rate initially. Most of the curves were nearly linear, but some of them came over. There was a definite fall in rate with one of these steels at a later stage. When it had reached a certain point we scraped off the corrosion products, and the curve started at the ordinary rate again and came over again.

Some analyses were made of the corrosion products, and it was clear that on some of the more resistant steels we were getting an extremely adherent layer of rust next to the metal, and that seemed to contain most of the Cr. The Cr tended to be enriched in this layer next to the metal. That bears on the point raised by Dr. Hoar. There is no doubt at all that usually with these low-alloy steels resistance is increased as a result of the formation of a very compact corrosion product.

Now, for what purposes would low-alloy steels be most beneficial if they do depend on a certain amount of rust before they develop their increased resistance? If a structure is such that it cannot be painted, clearly there is an advantage. But suppose it is one that will have to be painted at intervals, does the use of a low-alloy steel increase the intervals between painting? If so, it is valuable. If it does not it may be that in some cases, although the steels are more resistant, it may not be worth while using them.

Nowadays, apart from painted steels, we may consider steels where cathodic protection may be applied. There again one would imagine that there might not be any advantage in using a low-alloy steel in a condition where it is known that cathodic protection will be applied, although again there are theoretical considerations. It might be expected that, as in the absence of cathodic protection the corrosion rate would be less than with the ordinary steel, less current would be required to polarize the steel to the required potential. It might therefore still be worthwhile using a low-alloy steel where it is to be cathodically protected. That is a point that might be looked into.

Another point that interests us at Teddington is the use of corrosion inhibitors in waters. If corrosion inhibitors are to be used, is it any advantage to have a low-alloy steel to begin with? This is a difficult question to answer.

Finally, Cu does occur in steel and one can usually be sure of getting a certain amount of Cu. But we had the experience recently of having a steel from ordinary production, and it was extremely low in Cu. When we reported this to the suppliers they were very surprised, but they said that while it was unusual it could occasionally happen. It is not safe, therefore, to rely on say, 0·1% Cu in the steel. One might find occasionally that there was not. It is important that if we wish to use a steel where we need a little Cu we should not merely rely on the probability that it is almost certain to contain Cu, but we should make sure that it does contain Cu in the amount required for the purpose.

Dr. **W. H. J. Vernon**, O.B.E.: The outstanding point in these researches is the importance of the alloying elements, coupled with the presence of Cu in most of the steels employed. My remarks will refer only to the influence of Cu.

The agreement of the authors' results with the historic results of Buck (U.S.A.), as shown by the disposition of the curves in Fig. 4, is most gratifying. Here the results are presumably based on field tests, and not on laboratory (spray) tests. These, the authors state, were carried out according to the method of Swinden and Stevenson (published 1940); but they do not appear to mention the composition of the spray used in the tests.

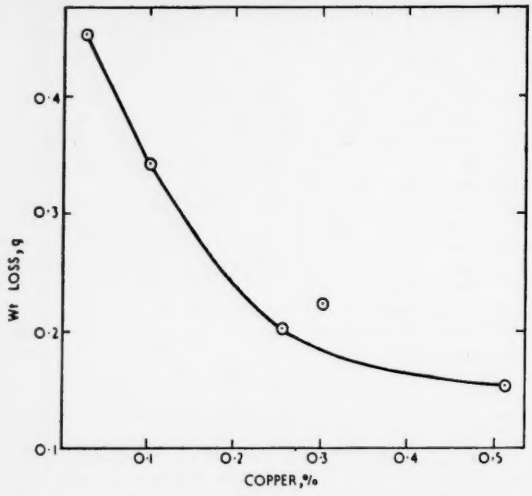

Fig. A—Curve of results obtained with spray based on sulphur-dioxide solution

In their original work, Swinden and Stevenson used a spray based on dilute sulphuric acid and sodium chloride. Some hitherto unpublished work was carried out at about the same time with Mr. E. G. Stroud, at Chemical Research Laboratory, Teddington. The apparatus used in this work was practically identical with the Swinden-Stevenson apparatus, for the very good reason that it had been, in fact, presented to us. Apart from a few structural modifications, the main difference was in the spray solution used. We had available a short series of Cu steels, from 0·2–0·5% Cu, and in a large number of spray tests using the sulphuric-acid–sodium-chloride spray we could get no correlation with the Buck curve. When, however, we substituted a spray based on sulphur dioxide in solution, together with ammonium sulphate (which in the intermittent intervals of drying out produces nuclei similar to those known to operate under service conditions) we obtained curves that consistently followed the general form of Buck's original curve. A typical set of results is reproduced in Fig. *A*.

It will here be observed that the point for the steel containing 0·3% Cu lies above the smooth curve. As the same thing happened in a large number of replicate experiments it was clear that this steel, which in fact had been obtained from a different source from the other steels, was behaving anomalously. A metallographic examination of the whole series revealed that whereas all had been annealed, only the 0·3% Cu steel carried its normal C content (0·15%): all the rest had been *completely decarburized*. This result bears upon the authors' reference to the influence of carbon except that it appears to work in the opposite direction. The authors found that low-carbon steels had a somewhat higher corrodibility than the steels with medium carbon content. On the other hand, this particular series happened to be straight C steels with virtually no Cu. It would therefore be of interest to know if any information is available from their field tests on the influence of carbon on truly Cu-bearing steels.

To complete the story of the experiments quoted I should add that, unfortunately, although the spray tests reproduced so excellently the service behaviour of straight Cu steels, this did not obtain for low-alloy steels containing Cr and other elements. A closer correspondence was obtained from various dipping treatments conjointly with the spray but the matter was not pursued because of the intervention of then more urgent problems. Dr. Wormwell has already referred to the question of pitting, and I may add that in these spray tests we found with the low-alloy steels a tendency to pitting that was not present with the straight Cu-bearing steels.

Finally, as one who for years has been preaching the advantages of Cu-bearing steels and urging engineers to use them for structural purposes, I should like to express my concern that these steels still appear to be unobtainable as such, i.e. except in so far as Cu may be present in sufficient amount purely fortuitously. Could manufacturers say whether it is possible for a buyer to specify, say, 0·25% Cu steel and to get it? In the absence of such assurance, advocacy of these steels on the grounds of increased longevity of structures appears to be useless.

CORRESPONDENCE

Mr. **W. G. Beynon** (for the Director of Naval Construction, Admiralty) wrote: This paper deals with the normal corrosion behaviour of low-alloy and other steels of various compositions. There is, however, a case of

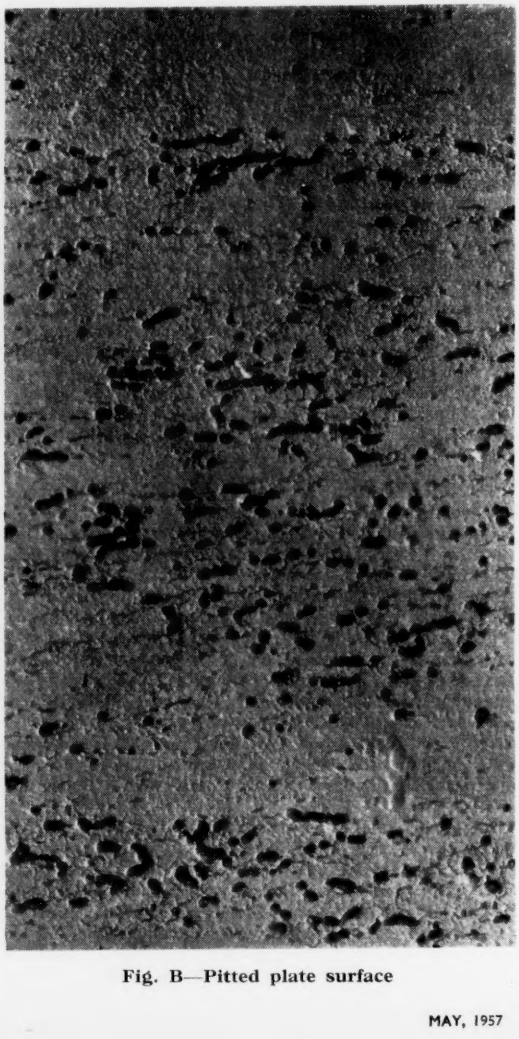

Fig. B—Pitted plate surface

abnormal corrosion behaviour in a low-alloy steel of the Mn–Mo type which is worthy of mention and in which considerable pitting occurred during storage in the open after pickling.

It does appear that the Mo-bearing steels tend to have a poor surface after hot rolling and it is suggested that this rough surface (Fig. *B*) entraps pickling fluids which remain after the normal washing. Such entrapped fluids and residues may be the cause of pitting over a period of years.

This behaviour has not been experienced with mild steel and steels of the C–Mn type.

AUTHORS' WRITTEN REPLIES

The **Authors** wrote in reply: We agree with Mr. Turner that additions of 3% of Cr or Ni are somewhat outside the practical range for the general run of structural steels, particularly as some steels of these types might require heat treatment or careful control of the carbon content, to avoid air-hardening.[2] In a limited number of tests made by ourselves in an industrial atmosphere, variations in Mn content of up to 1·5% did not affect the corrosion resistance of steel materially, but there is evidence from U.S. sources that the Mn content may have an effect in marine atmospheres. The British Iron and Steel Federation are considering the possibility of suggesting to the British Standards Institution that reference might be made in appropriate specifications to the superior resistance of low-alloy steels to atmospheric corrosion. Moreover, as Mr. Turner will be aware, the Atmospheric Corrosion Sub-Committee has drawn attention to the Atmospheric Corrodibility Index in a letter to the technical press.[3]

Dr. Main rightly points out that the systematic investigation of the corrosion of ferrous metals under marine conditions undertaken by the Institution of Civil Engineers preceded the researches of the Corrosion Committee of The Iron and Steel Institute by a number of years. The former tests yielded valuable results for a representative range of irons and steels, such as were available at the time, although they naturally did not cover materials that have been introduced more recently.

Of the two mechanisms suggested by Dr. Hoar for the superior resistance of low-alloy steels to atmospheric corrosion, we regard the second as by far the more important. Potential measurements made by Dr. F. Wormwell and others have shown that, although low-alloy steels are 'noble' to ordinary steel free from alloying elements, the potential differences between the two types of steel are relatively small; in any case they would rarely come into play under atmospheric conditions. Possibly these potential differences might be turned to account under other conditions, as, for example, by using low-alloy steel rivets for ships' hulls, when the rivets, in theory, would be protected by the ordinary mild steel plates. However, for resistance to atmospheric attack, the low-alloy additions operate mainly by their influence on the rust. As shown in Fig. *C*, there is a direct correlation between the resistance of a steel to atmospheric corrosion and the sulphate content of its rust, which is greater for low-alloy steels than for ordinary steel; similar observations have been made by H. R. Copson in the U.S. Some work of Schikorr is also of interest in this connection. He showed[4] that whereas non-ferrous metals as a class are generally converted almost entirely into sulphates by atmospheric corrosion (in an urban atmosphere), in the case of Fe only a small proportion of iron sulphate is formed, which apparently acts as a catalyst for the rusting process. It seems, therefore, that when low-alloy steels rust in the open air, the small amounts of non-ferrous metals present in them are converted into sulphates. These sulphates are

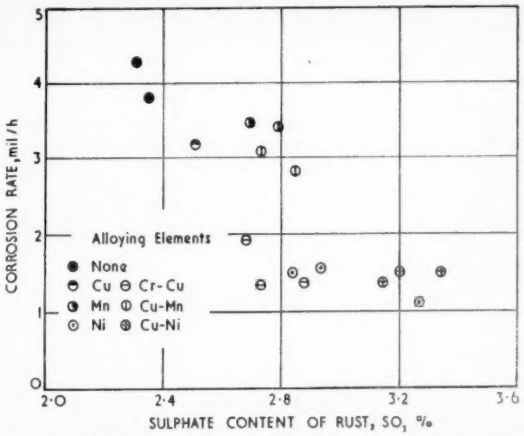

Fig. C—Diagram showing direct correlation between resistance to atmospheric corrosion and sulphate content of rust

hydrolysed to basic sulphates and gradually accumulate in the rust on which they confer a much more compact and impermeable texture. This theory is supported by the fact that where, as in railway tunnels, the natural leaching processes associated with rain and weather cannot come into play, there is no appreciable difference between the corrosion rates of ordinary and low-alloy steels.

In reply to Dr. Wormwell, we confirm that under our conditions of exposure in sea-water, in which the corrosion products remained in contact with the steel and formed a continuous layer on them, there was very little pitting and certainly none that could be attributed to the compositions of the steels themselves. We anticipate that the benefits accruing from the use of low-alloy steels for painted atmospheric structures will arise not so much from a lengthening of the optimum painting interval, as in the retardation of the corrosion process when this interval is exceeded, as is often the case. To illustrate this, some years ago the painting schemes on stands of low-alloy steel at the Corrosion Committee's Sheffield exposure station were found to be in much better condition than on stands of ordinary steel treated in the same ways.[5] Now, 17 years later, it has become necessary to remove these stands to another site, after all the vicissitudes of the World War II which led to considerable damage to the paint, and, it must be admitted, some falling short of the ideals for frequency of maintenance painting. The superiority of the low-alloy steel stands has become very noticeable, for they have suffered appreciably less corrosion than the ordinary steel stands.

Dr. Vernon's results for laboratory tests on a series of Cu-bearing steels are an interesting addition to our own data. Although we cannot be certain that the C content of such steels has no effect on their resistance to atmospheric corrosion, we should attach greater importance to possible variations in S content, such as, for example, have sufficed to explain an apparent discrepancy between our results and those of Buck.*

In reply to Dr. Vernon's question, the spray used in the laboratory tests discussed in the present paper was a solution that was N/100 with respect to both sulphuric acid and sodium chloride. Swinden and Stevenson were originally led to substitute this type of spray for the more commonly used sodium chloride solution because the main constituents of atmospheric pollution in this country are sulphates rather than chlorides.

* *See* p. 280, l.h. col., and Fig. 4 of the paper.

We have no knowledge of the effects of Mo to which Mr. Beynon refers, but there is no doubt that certain alloying elements affect the surface condition of hot-rolled steel and, in consequence, its resistance to corrosion and pitting. For example, globules of Cu have been found in the millscale on Cu-bearing steel and care is necessary in rolling such steel to avoid roughening at the edges of sections. The practical remedy in the case of ship plate would seem to lie in removing the millscale by blasting instead of by pickling in an acid bath.

Finally, we should like to thank all who have contributed to the discussion, and we reiterate that although we appear as authors of this paper, the work is really the result of a combined effort in which we have been materially assisted by many friends.

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CORROSION RESISTANCE OF WROUGHT IRON

This discussion was based on the paper **"The Corrosion Resistance of Wrought Iron"**, *by J. P. Chilton and U. R. Evans*, which appeared in the October, 1955, issue of the *Journal* (vol. 181, pp. 113–122).

Dr. **U. R. Evans** (Dept. of Metallurgy, University of Cambridge) presented the paper.

Dr. **F. Wormwell** (Chemical Research Laboratory, Teddington): We all recognize that the subject investigated by Dr. Chilton and Dr. Evans is one of great interest, but it is one that has clearly presented considerable difficulties. The authors are to be congratulated on giving a reasoned discussion of the factors that contribute to the properties of wrought iron concerned with its resistance to corrosion.

Dr. Evans referred to the great variability, and this prompts me to recall that in some work done with Dr. Bengough many years ago, where from samples of wrought-iron rod that was piled, we cut discs and exposed them. On the face of the disc we were exposing the wrought iron corroded between the zones, although at the edges the zones were perpendicular to the corrosion and might have retarded corrosion. But we did find the corrosion/time curves were slightly more irregular than those for mild steel, and the method was sufficiently sensitive to show little wobbles from day to day in the corrosion rate, which indicated that the material was variable.

Dr. Evans has given a closely reasoned account of why these variations occur and why in some circumstances wrought iron can be very much more resistant than mild steel. But there are still some puzzling questions to be answered, and when deciding whether to carry out the further considerable research that would be required, the question arises whether it would be justified in relation, say, to the development and application of low-alloy steels. That is a point Dr. Evans touched upon, and it is a matter that might very well be discussed here.

It is still not quite clear to me why the Q and R zones should alternate in such a regular manner. It reminds one of periodic precipitation phenomena, but surely it is not produced by the same effect. Perhaps Dr. Evans could make some observations on that. One can understand that the material is inhomogeneous, but why should it come out in such neatly arranged alternate layers of Q and R zones?

The general conclusions appear to be based on corrosion experiments in acid solutions, on anodic corrosion experiments, and on the results of very elegant micro-analysis by Berger. I wonder whether any long-time corrosion experiments were carried out either in atmospheric conditions or by immersion in neutral solutions, apart from the examples illustrated in Figs. 9 and 10.

In connection with Fig. 9, would Dr. Evans consider that a content of about 3% nickel as in the V zones and 0·75% copper as reported in Table III (the highest figure actually obtained in the analyses) would account for the complete absence of rusting, apparently, in the V zones?

In introducing the paper, Dr. Evans said that the actual content might well be higher. One would imagine that if a zone resists rusting completely for any period in a corrosive environment, it must contain more than those amounts. It is known that much higher contents of nickel and also of chromium would normally be required to render a steel completely unattacked in most atmospheres.

The possibility occurs to me that there might be some form of cathodic protection. The conditions of exposure might be such as to permit of a conducting path with some dissolved salts, which would mean the corrosion of the adjacent R zones was cathodically protecting the V zones or helping in that way.

Dr. Evans mentioned a very elegant technique of adding hydrogen sulphide to the liquid to see what happened, but I do not think he says what concentration was used. No doubt it is difficult to say whether the concentration of hydrogen sulphide used in the liquid is comparable with the concentration of hydrogen sulphide that could be expected to be coming from the steel. The results would undoubtedly support the theory that the susceptibility of the steel to attack is controlled by the amount of available sulphur. But one could imagine the possibility that swamping the liquid with sulphide might give the result obtained, even if the original cause of the difference was nothing to do with sulphur. It is a matter of logical deduction, but the probability is, from what Dr. Evans has said, that other supporting evidence all tends to support the idea that the sulphur is really the important factor.

It seems from the paper that the reasons for the resistance of wrought iron in certain circumstances might be (1) the presence of a resistant scale produced during manufacture; (2) the existence of R zones; (3) the existence of V zones. Would Dr. Evans agree that when structures of any size are virtually uncorroded (i.e. they persist for very long period) this is attributable to the presence of a good protective scale combined with a low probability factor for corrosion without breaks in the scale? Probability considerations alone could hardly explain the virtual absence of corrosion over very large areas.

A final point is that when wrought iron shows not complete resistance, but a high resistance relative to mild steel, this result might be brought about by a

combination of the properties of the *V* and *R* zones, thanks to the material itself and then to the environment in which it is placed, including the constituents of the water. In some unpublished work at Teddington we have obtained some evidence with certain low-alloy steels that they show up to better advantage in waters containing calcium bicarbonate. The good result obtained is due to a combination of the influence of the alloying element in the steel and the presence of certain constituents in the water.

Dr. **W. H. J. Vernon,** O.B.E.: One of the illustrations that figures prominently in the paper is a rubber stamp. A 'rubber stamp research' is surely the last thing one would expect from Cambridge! This, however, is only one of the ingenious devices used by the authors, and indeed by many other researchers in Dr. Evans' laboratory, which gives us important results with the minimum of elaborate experimental techniques.

Other speakers have dealt very ably with detail. I should like myself to emphasize that this investigation has clarified a subject on which there has been over many years a good deal of obscurity, and has provided a scientific basis for what was previously very largely empirical or conjectural. If, as we have been told, wrought iron is going out of production, it is just as well that we shall understand the mechanism of whatever superiority in corrosion resistance it may possess before that event comes to pass. The knowledge will assuredly not be lost.

The authors in their introduction mention rather casually the old barrier theory of the corrosion resistance of wrought iron, in which attack was considered to be impeded by parallel layers or flakes of slag. That theory was, in fact, introduced by the American metallurgist, Prof. H. M. Howe, just 50 years ago, and has since been much used to present to students a picture which is perhaps all too convincing. As the authors point out, the fallacy is exposed by considering, instead of flat bars of wrought iron, rods of circular section in which the slag will take the form of strings or threads.

The Cambridge work has done a great service in deflecting emphasis from the slag, as such, to the contiguous metal; and although the authors point out that the idea of zones in the metal phase is not new, the more critical approach has led to the demonstration, as shown in previous papers, of the existence and function of the *Q* and *R* zones, and now, in the present paper, the *V* zones. I had thought that Dr. Evans and his associates, having marked out the *Q*, *R*, and *V* zones, might have 'no more territorial ambitions'; but from Dr. Evans' earlier remarks it looks as though there may be yet another (possibly the '*VV*' zone) of considerable practical importance.

The immediate question is, how can these findings be implemented; how can the knowledge be applied to the benefit of the engineer of the future? I am tempted to revert to the preceding discussion, in which Mr. Turner mentioned the occasional inadvertent enrichment of copper in the fabricated surface of copper-bearing steel. One wonders whether this capricious enrichment, sometimes apparently leading to too much copper, might not be brought under control, possibly by selective oxidation and a suitable pickling treatment, with a view to producing what would virtually be a *V* zone at the metal surface.

In fairness to the sponsors of the present research it should be stated that the conditions under which an established greater longevity of wrought iron might be held to justify its continued use for boiler tubes, were those of trawlers at sea, where controlled water treatment was normally ruled out.

Dr. **T. P. Hoar** (Department of Metallurgy, University of Cambridge): Dr. Evans' remarks on the probable effect of sulphur in these heterogeneous wrought iron results are interesting.

Many years ago I studied the corrosion resistance of the then usual steel used for tinplate, a hot-rolled semi-rimming steel. The surface layers contained considerably less sulphur than the inner parts. On a cut sheet corroding in weak acid, the surface layers at the edge were sometimes noticeably less corroded than the middle zone, an effect very similar to that found in wrought iron producing the *Q* and *R* zones.

In the work to which I refer, we examined the effect of an added excess of H_2S on the rate of corrosion of steels in acid. Such additions caused low-sulphur, slow-corroding steels to corrode as rapidly as high-sulphur, rapid-corroding steels. Conversely, small amounts of tin or cadmium ions, which inhibit the corrosion of mild steel in acid solutions, reduced the corrosion rate of the fast-corroding steels to that of the slowly corroding steels: tin and cadmium sulphides are not soluble in acids, and consequently H_2S was sequestrated. Have Dr. Evans and Dr. Chilton examined the influence of tin and cadmium additions on the acid corrosion of wrought iron in weak acids? It would be interesting to see if the *Q* and *R* zones become less distinct under such conditions.

Dr. **D. Eurof Davies** (University College, Swansea): There has been an apparent inconsistency in the literature as far as the corrosion resistance of wrought iron and mild steel are concerned. After reading this paper one can understand how these inconsistencies have arisen, and why it is that under certain circumstances, wrought iron does behave in a better way than mild steel as far as corrosion resistance is concerned.

It seems that sulphur plays an important part in the corrosion of wrought iron and the authors have set out their theory in a very convincing way. Whether a particular zone is attacked or not depends on the accessibility of the sulphur, and the authors state that when the sulphide is enclosed in an envelope of slag as in the *R* zone, that zone is not attacked, whereas in the *Q* zone the sulphide inclusion is not enclosed in slag and hydrogen sulphide is therefore liberated. In what form is the sulphur? Is it present as the sulphide of iron? The analysis shows that there is some manganese present and although the amount is small, could the inclusions be manganese sulphide or possibly iron manganese sulphide?

What evidence is there, apart from Collari's work, concerning the films of slag around the sulphide inclusion. Do the authors envisage a sulphide inclusion inside a thin shell of slag, or is the inclusion surrounded by a relatively thick layer of slag? Have the authors investigated the complex structures of the slag layers found in wrought iron, and in what way are these related to the sulphide inclusions mentioned in the paper?

Like Dr. Wormwell, I find the regularity of the *Q* and *R* zones very puzzling. It seems that the only way of getting at the root of the matter is to examine the wrought iron as it is being produced, i.e. during different stages of manufacture. It seems that these zones are tied up with the rolling of the wrought iron ball, and it would be interesting to find out when, during the rolling, these various *Q* and *R* zones manifest themselves.

Wrought iron has now been in use for a few thousand years, but is dying out. Before it finally passes from the scene it would be interesting to have a complete picture of the uniformity of the *Q* and *R* zones.

Dr. **G. Butler** (Chemical Research Laboratory, Teddington): Like other speakers, I am a little puzzled by the mechanism by which the sulphur becomes inaccessible.

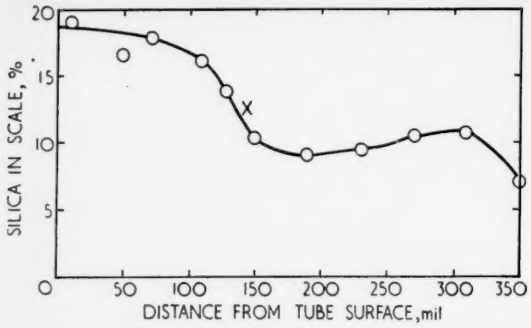


Fig. A—Silica content in wrought-iron boiler tubes

If it is imagined that the sulphur is contained in, say, a slag envelope it would be expected that the length of the resistant lines shown in the etched cross-section would be similar to the length of the slag inclusions. My impression is that the former are much longer.

Turning to the low figure for sulphur in the table on p. 117, what evidence is there that the total sulphur was not lower in this zone? The accessibility was tested by reaction with 6N hydrochloric acid. Were any checks carried out using more vigorous solution techniques to determine whether the sulphur was actually present? More data on the sulphur accessibility would be desirable particularly since on the same page it is said that great inhomogeneity exists within the zones. Could this account for the variation in the sulphur content given in the table?

An interesting point has arisen in our own work on boiler corrosion which is being done for the British Shipbuilding Research Association. The thick scale formed on a wrought-iron tube which had been in use for about 40 years was examined. The tube was mounted in the lathe and 20-mil turnings were taken and analysed for silica. In Fig. *A* the silica content of these samples is plotted against the distance from the tube surface. A marked increase in concentration is obtained on the tube surface. Owing to the slightly eccentric mounting of the tube in the lathe, iron was already exposed to the left of point X. There is good reason to believe that, had the tube been mounted centrally the silica concentration at the surface would be found to be even higher. This strongly suggests that the corrosion resistance of wrought iron may be linked with the formation of a very adherent impervious scale by the combination of the silica in the wrought iron and corrosion products and scale deposits.

Mr. **T. Henry Turner** (British Railways Research Department): This paper adds to our knowledge, but one cannot envisage a large increase in the use of wrought iron in the U.K. For chains it must be almost dying out. From laboratory tests of mild steel chain compared with wrought iron chain, there is little to justify the use of wrought iron, even if it were available.

It is difficult to imagine a chain formed with *V* zones or even the *R* zones spread all the way over the outside of the chain link.

Even if we now know about why wrought iron corrodes and how it corrodes, it seems to be a more practical consideration that the French Railways now have in use some 1300 steel-tubed and steel-fireboxed locomotive boilers which have not had a tube changed, and the fireboxes are still the same after ten years' service of about 1 million kilometres running. That real revolution in engineering practice was produced not by changing the nature of the metals that are used, but by faith and discipline in water treatment changing the nature of the environment.

The importance of environment can be seen, as the authors have no doubt observed, by studying the wrought iron gates and railings in Oxford and in Cambridge. One may assume that the wrought iron has the same number of *Q*, *R*, and *V* zones in its make-up in either city, but wrought iron is very much more corroded in Oxford than in Cambridge; the environment has had the outstandingly important effect.

AUTHORS' REPLIES

Dr. **Evans:** I should like to thank all those who have taken part in the very admirable discussion.

I was particularly glad to have Dr. Wormwell's agreement about the variability of wrought iron, both in the sense in which I had used the word 'variability' and in a slightly different sense, namely, the wobbles in the corrosion time curve.

As he says, there are a number of puzzling features remaining to be solved. I think we have shown why *R* zones are more resistant than *Q* zones, but we have not shown why *R* and *Q* zones exist. He referred to the very uniform distribution of *R* and *Q* zones. I am not sure whether Dr. Chilton would entirely agree that they were uniformly spaced out in the iron used by him. *V* zones, of course, should be uniformly spaced, providing that the components of the pile are of the same thickness. It is my impression that in practice *V* zones are rather 'chancy', some broad, some narrow, and not spaced at a uniform distance from one another.

As to why they occur longitudinally, various speakers have rightly pointed out that this results from rolling. I am not a rolling expert, but I imagine that if you take an ingot with different regions having different mechanical properties and roll it out, there will inevitably be layers of the kind met with in wrought iron.

As to whether 3% nickel is sufficient to account for the absence of rusting, I think (as, indeed, Dr. Wormwell has pointed out) that the true nickel content at the exact centre of the *V* zone would probably be much higher. I also agree (as mentioned in the paper on p. 118 in connection with *R* zones) that the intense corrosion in *Q* zones would give cathodic protection to neighbouring *R* zones, and also to *V* zones.

I am inclined to agree with Dr. Wormwell's summing up as to the good behaviour of certain large-scale structures, but I would repeat here what is stated in the paper (p. 113), that the samples of wrought-iron structures which we have to-day are probably not typical of the population. The good ones have survived, and the bad ones have departed, so that the result is perhaps unduly favourable.

I am glad that Dr. Davies is here. The two-phase structure of the slag constituent has been brought out clearly. When Lewis was working at Cambridge in the middle thirties he showed me microsections of the slag inclusions, displaying a complex structure. I cannot quite remember whether Lewis made these himself or whether he found them in an earlier publication, but I do remember that they showed at least two phases; sometimes there appeared to be more than two. There was a tendency for one phase to be inside the other, presumably as a result of interfacial energy relationships. The phases will always tend to arrange themselves in such a way as will reduce the total interfacial energy of the system, and this will often result in one phase enveloping the other; examples are known in non-ferrous metallurgy.

I was particularly glad that Dr. Butler was able to contribute to the discussion, since it gives me an opportunity of expressing my appreciation of the permission to reproduce his very instructive photographs (Fig. 11).

As regards the question whether sulphur was really present in the *Q* and *R* zones, I think there is no doubt that the sulphur contents of the *Q* and *R* zones are not so very different. This is shown in Table II (p. 116). The difference comes out in the comparison between the fine and coarse millings on p. 117.

Dr. Hoar's story of the ingot has reminded me of something which occurred about the same time. I am not sure whether it refers to the same ingot or a different one. At that time Dr. Mears was working on the probability of corrosion, and we had amongst other things a small ingot of free-cutting steel with a sulphur segregate in the middle. It was shaped rather like a long loaf of continental bread, and every now and then Mears cut off a slice for his experiments, and the remainder of the ingot, with a new bright face, was laid up against the wall. After a time the centre, where there was sulphur, developed rust and the rim remained perfectly bright.

Dr. **Hoar:** The one I drew was thin plate steel.

Dr. **Evans:** That would doubtless show the same sort of effect. We have not tried tin or cadmium, but I fully agree with the explanation given by Dr. Hoar.

I was very interested in Mr. Turner's remarks, and I believe that the early history of this research which he gave is the correct one; Dr. Pearson knows the details better than I do. The case quoted where correct water treatment on the French Railways caused tubes to last ten years without deterioration represents a very important achievement and I hope that details will be given when the remarks are recorded.

I would particularly thank Dr. Vernon for his kind remarks, and I am glad that he called attention to Dr. Chilton's rubber stamp. As a matter of fact, we did not use the stamp much in this research, but it would have been very useful for Dr. Mears' work had it been available in 1935. It was entirely Dr. Chilton's invention.

The slag barrier theory mentioned by Dr. Vernon was one which also appealed to me, and I share his regret that it must go! I held rather similar views in 1928 in the first of the three papers from Cambridge (the second was by Lewis and the third by Chilton). I then used the theory to explain what may be called the 'dry dock effect'. I coated a piece of wrought iron with tar-paint then engraved a scratch line exposing the metal. The specimen was subjected to anodic attack so as to eat a groove into the metal; on wrought iron this gave a stepped outline like the sides of a dry dock, whereas on steel the section was more or less semi-circular.

At that time the result seemed to be due to the attack on the wrought iron coming up against barriers and being diverted sideways. The true explanation, I think, is a little different. As long as the attack passes through a *Q* zone it can proceed downwards. When it comes to an *R* zone the absence of accessible sulphur as an anodic catalyst is such that it slows down in the *R* zone and proceeds sideways along the *Q* zone. Then attack on the *Q* zone gives cathodic protection to the *R* zone. The new theory is not really very different from the old barrier theory, only instead of a mechanical barrier we have an electrochemical barrier.

FUEL-ASH CORROSION OF STEELS AND HIGH-NICKEL ALLOYS

This discussion was based on the paper by H. T. Shirley entitled **"Effects of Sulphate–Chloride Mixtures in Fuel-Ash Corrosion of Steels and High-Nickel Alloys"** which appeared in the February, 1956 issue of the *Journal* (vol. 182, pp. 144–153).

Mr. **H. T. Shirley** (Brown-Firth Research Laboratories) presented his paper.

Mr. **P. H. Crumley** (Fuel Research Station, Greenwich): My particular interest is in boiler availability, i.e. deposits and corrosion of boiler surfaces. Conditions in boilers are less arduous than conditions in gas turbines with which Mr. Shirley is concerned. He is interested in high alloys at temperatures around 750° C, while we deal with metal temperatures of about 500–600° C and fairly low alloys.

I should like to ask Mr. Shirley for his views on superheater tube wastage, which although not as serious as cold-end corrosion in boilers, is by no means uncommon. When it does occur it often has an obvious cause, such as steam starvation, but we know of one particular case where superheater tube wastage occurred from the firing of high-chlorine coal.

The boiler had a steam temperature of about 850° F, but the metal temperatures were rather higher than would be expected from the steam temperature. The coal fired had a chlorine content of about 1% and tube failures occurred much more quickly than would be expected. There was considerable thinning of the metal wall at the point of the burst, and the deposits contained significant amounts of chloride. Does Mr. Shirley think that the tube wastage could be the result of chloride attack?

The calcium content in the ash of this particular coal was of the same order as the peat in the analysis given on p. 148, which is rather high for coal. Table VI shows steels were more rapidly attacked by the calcium than by the sodium, so it may be possible that the high calcium content of the ash was in this instance partly responsible for the rapid attack.

It is not at all clear why peat ash should be more corrosive than coal ash. There may be some mineralogical difference, or the ash analysis may be relatively unimportant but the sulphur and the chlorine in the fuel may be important. Chlorine is fairly low in peat, but the sulphur is very low, so that the sulphur/chlorine ratio might be dangerous.

At the Fuel Research Station some work has been done on the conversion of sodium chloride to sodium sulphate. We were primarily interested in deposit formation, but the findings may be of interest with respect to corrosion. The first experiments were with the burning of kerosene doped with sodium naphthanate, carbon tetrachloride, and carbon disulphide. We determined the amount of chloride and sulphate in the flue gases with different sulphur/chlorine ratios, and found that with a sulphur/chlorine ratio of more than about 4 : 1, the amount of chloride was very small. However, we could never entirely eliminate the chloride, so that possibly even a high sulphur/chlorine ratio will still cause some corrosion.

Similar work was done adding sodium chloride to the kerosene and again the chloride was mostly disposed of with quite moderate sulphur/chlorine ratios, but we could not eliminate it all.

Work was also done on the formation of sodium sulphate

from sodium chloride in conditions similar to those on boiler tubes, using a radioactive tracer technique, and has been reported by Fletcher and Gibson.* They heated sodium chloride in the presence of SO_2 and SO_3. In the absence of a catalyst they found it nearly all came from SO_3 at temperatures up to 600° C, but above that temperature the SO_2 also reacted. In the presence of a catalyst it nearly all came from SO_2.

Finally, I believe that Mr. Shirley has tried the effect of adding carbon graphite to his test mixes to simulate the effect of unburnt carbon, and he found it did not have very much effect on corrosion. From that it may be concluded that bad combustion is not a very important factor in corrosion. Going back to his 1950 paper, Mr. Shirley showed there that a badly mixed stoichometric mixture of gas and air leads to more corrosion that a well-mixed gas mixture with insufficient air. Therefore I wonder whether adding graphite is a fair imitation of poor combustion, for it is very well known that bad combustion will give a rapid tube wastage at quite low temperatures.

Mr. **Shirley:** The reason for the tests with graphite was the reported presence of unburned carbonaceous matter depositing with the ash during gas turbine trials. These tests must be taken to refer only to this aspect and not to the wider general field of poor combustion.

Dr. **T. P. Hoar** (Department of Metallurgy, University of Cambridge): Mr. Shirley's comparative figures for the different high-alloy steels and alloys are interesting, particularly those showing the effect of chromium. It has been suggested elsewhere that useful protection of high-alloy steels and of Ni-based alloys can be achieved by chromizing the surface by one of the high temperature chromizing processes that produce a high-chromium layer. Would Mr. Shirley care to comment on this?

Are there any comparative data on materials other than the commercial alloys mentioned in the paper? It would be most interesting to see the effect of sulphate on a series of pure Fe–Cr and Fe–Ni alloys.

Dr. **K. Sachs** (G.K.N. Group Research Laboratories): Like previous speakers, I should like to express my appreciation of this excellent paper. Some years ago I was associated with the investigations at the Research Laboratory of the Mond Nickel Company into the peculiar 'warts' found on some Ni–Fe–Cr alloys exposed in experimental gas turbines. In this context I should like to point out that our experiments which identified the light grey phase as chromium sulphide have been published,† presumably since Mr. Shirley's paper was completed.

The effect of traces of chloride in promoting sulphur attack raises a number of interesting points. Sulphide penetration in nickel-rich alloys is not unknown but is generally confined to reducing atmospheres; moreover, nickel sulphide is formed rather than chromium sulphide under these conditions. In the presence of chloride, chromium sulphide is formed, and the phenomenon is observed over a range of atmospheres. Mr. Shirley reports experiments on the effect of different atmospheres and of small particles of graphite. Has the effect of vanadium pentoxide been related to the chloride effect? Has it been shown that the chromium sulphide attack occurs in mixtures containing not only sulphate and chloride but also vanadium pentoxide? Would it be reasonable to postulate that vanadium pentoxide has the effect of further lowering the chloride concentration at which this particular kind of attack may occur?

Such a postulate might help to explain the effect of sulphates in making nickel-based alloys susceptible to vanadium pentoxide attack. In the presence of pure vanadium pentoxide, iron-rich austenitic heat-resistant alloys suffer catastrophic oxidation to which nickel-base alloys are substantially resistant. However, mixtures of vanadium pentoxide and sodium sulphate have much more effect on nickel-base alloys than pure vanadium pentoxide. Here again, a sulphur-bearing compound is exercising a deleterious influence on nickel-base alloys in the absence of reducing atmosphere, indeed under conditions made more oxidizing by the presence of vanadium pentoxide. If, as has been postulated, vanadium pentoxide enables local penetration by chromium sulphide to proceed at very low chloride concentrations, the effect of mixtures of vanadium pentoxide and sodium sulphate on nickel-base alloys can be attributed to very small traces of chloride.

AUTHOR'S REPLY

Mr. **Shirley** (Brown-Firth Research Laboratories): I was most interested in Mr. Crumley's remarks, and the work carried out at the Fuel Research Station on chloride-sulphate conversion is of special importance.

Examination of deposits from service tubes does not necessarily give the initial composition of deposits because of continued reactions with other combustion products, leading, for example, to almost complete sulphation of chlorides. In this connection it is very desirable that efforts should be made to obtain accurate indications of the nature of the initial deposits through development of probe techniques.

Primary deposits are probably very variable. It has been found that material collected from a cyclone separator fitted prior to the gas turbine in a peat set contained quite considerable amounts of unsulphated oxides, so obviously sulphation does not all occur in the flame zone or even in the primary settling zones, and it appears that varied deposits may occur more or less by chance contact between the surfaces and particular constituents of the flue gas stream. The action of these deposits on metal components will very largely depend upon protection afforded by earlier deposits, including the degree of their removal by erosion.

This has relevence to Mr. Crumley's point about superheater tubes. Generally speaking, however, coal ash is much less dangerous than peat or oil ash. This may have a mineralogical explanation, dangerous constituents being less readily available in the coal ash. With the generally lower temperatures of steam practice there has been relatively little trouble from ash attack in the past. In my experience with carbon or low-alloy steel superheater tubes, major attack has always been associated with deposit removal by steam lancing. Sometimes in such places there has been very severe attack, but because of the erosion no evidence remains as to whether or not corrosive deposits may have contributed. On the other hand, if temperatures rise above 600° C, more direct ash attack can certainly be expected.

Dr. Hoar has mentioned high-chromium alloys. Plain high-chromium alloys have shown surprisingly high resistance under complicated service conditions involving vanadium as well as sulphate deposits. Whether or not advantage can be taken of such behaviour naturally depends upon requirements in relation to the physical and mechanical properties of the steel. With regard to Dr. Hoar's question on the behaviour of pure Fe–Cr and Fe–Ni alloys, we have no additional data. Our tests, apart from those on pure nickel and chromium, have been on normal commercial alloys.

Dr. Sachs raised the question of combined action of vanadium pentoxide with sulphates and chlorides. Tests

*A. W. Fletcher and E. J. Gibson: Proceedings of the Second Radio-isotope Conference, Oxford, 1954.

†*J. Inst. Petroleum*, 1955, vol. 41, no. 377, p. 170.

in oil-fired steam raising equipment have indicated severe and complicated effects under such conditions with heavy sulphide attack in spite of the oxidizing nature of V_2O_5. Microstructural effects are particularly complicated with formation of a series of zones as the primary attack moves into the metal.

Since writing the paper we have carried out some further tests with other than sodium and calcium salts. For example, we have tested chloride contaminated magnesium sulphate with Nimonic and in spite of rumours that this might be very troublesome, have found it similar but rather milder than sodium sulphate. We have also tested aluminium-sulphate–sodium-chloride mixtures with F.C.B.(T) and Nimonic at 750° C. Here the indications are that these mixtures behave more like calcium than sodium salts, but are much less dangerous.

It was felt that some tests with sodium–lithium salt mixtures would be of interest, as providing a fully molten mixture at 750° C to allow clearer study of the minimum percentage chloride required for the onset of severe attack. Adding sodium chloride, the attack became more severe at about 2% and was heavy at 5% addition. In relation to these threshold values it must always be remembered that in practice random or chance effects may considerably enhance local concentration above average.

I would like to draw attention to some work of great interest when considering these effects. In the U.S.A. Simons, Browning, and Liebhafsky* working with sodium sulphate, have produced strong evidence for the auto-catalytic nature of sulphide attack once some initial reduction of sulphate has occurred. Add to this the finding by Buckland, Gardener, and Sanders† of the severe effect of sweaty fingerprints in initiating catastrophic failure in synthetic ashes, and the potential importance of even extremely local concentrations of chloride is apparent.

* *Corrosion*, 1955, vol. 11, pp. 505t–513t.

† Paper A–52–161 of Annual Meeting of the American Society of Mechanical Engineers, 1952.

DISCUSSION ON STRESS-CORROSION CRACKING OF STAINLESS STEELS

This discussion was based on the paper **"The Stress Corrosion Cracking of Austenitic Stainless Steels"**, *by T. P. Hoar* and *J. G. Hines*, which was published in the *Journal* in two parts, Part I appearing in February, 1956 (vol. 182, pp. 124–143), and Part II in October, 1956 (vol. 184, pp. 166–172).

Dr. **T. P. Hoar** (Dept. of Metallurgy, University of Cambridge) presented Part I of the paper, and Dr. **J. G. Hines** (I.C.I. Ltd. Billingham Division) presented Part II.

Dr. **R. N. Parkins** (Dept. of Metallurgy, University of Durham): The authors are to be congratulated on a most thorough investigation in a field which is becoming increasingly important as the problem of general corrosion is overcome and the conditions of usage of metals moves to those borderline regions between corrosion and passivity.

The authors have interpreted their results as indicating that the stress-corrosion process is dominated by the initial or incubation period, during which the presence of stress is incidental. It is clear, however, that the relative lengths of time occupied by crack incubation and crack propagation must be a function of the thickness of the specimen being tested. Any autocatalytic reaction of the type envisaged by the authors must, by its very nature, require an appreciable period of time for the conditions for propagation to be achieved and, presumably, these conditions will be the same for thin or thick specimens. The crack propagation will begin when a precise chemical condition has been established in a flaw of some typical depth. In very thin wires of the type used in the work now being discussed, the depth of these flaws may constitute an appreciable part of the cross-section and, at the higher stresses at least, it may be that what is being observed would be more aptly described as mechanical rupture triggered off by corrosion pitting, rather than stress-corrosion cracking in the commonly accepted sense.

Herein may lie an alternative explanation to that offered by the authors for the difference in slope of the applied-stress/cracking-time curves at high and low stresses. Thus, instead of the suggestion that at lower stresses it takes longer to achieve the conditions for rapid propagation, the mechanism of cracking presumably remaining the same, it may be that the crack propagation process itself is quite different corresponding more closely to a true stress-corrosion process. The potential/time curves of Fig. 12 could be interpreted in such a manner.

Although the authors suggest that their specimens contained internal stress, they also claim that stress is not important during the incubation period. Some experiments* on mild steel, where the mechanism of stress-corrosion is obviously likely to be different from that now being discussed but where certain similarities also exist, seem worth recalling in this context. The experiments were concerned with testing carefully annealed material under three different conditions:

(i) With a stress applied from the beginning of the experiment
(ii) With cold worked material corroded for the time required for cracking in (i) before the same stress was applied
(iii) With annealed material corroded for the time required for cracking in (i) before the same stress was applied.

The total time of corrosion to the point when failure occurred was similar in tests (i) and (ii), but was twice as long in test (iii). This suggests that applied stress and internal stress behave in much the same way and to conclude that stress is unimportant during the early stages of the corrosion process may not be valid if the specimens contain internal stress, even though no applied stress is present.

Finally, it is inferred from the suggested mechanism that there is no metallographic aspect to transgranular stress-corrosion cracking in stainless steels. The photomicrograph Fig. *A* shows the appearance of transgranular stress-corrosion cracking in a piece of $\frac{3}{8}$-in. thick stainless steel plate very similar in composition to one used by the authors and tested under identical conditions. The relation of the crack path to the twin bands seems to be more than coincidental and would suggest that there is in fact some metallographic aspect to the mechanism. It would be interesting to have the authors' views as to whether the differences between the

* R. N. Parkins, "Stress Corrosion Cracking of Mild Steels", read at Boston Conference of the Electrochemical Society, 1954, and published in "Stress Corrosion Cracking and Embrittlement", 1956, Wiley, New York.

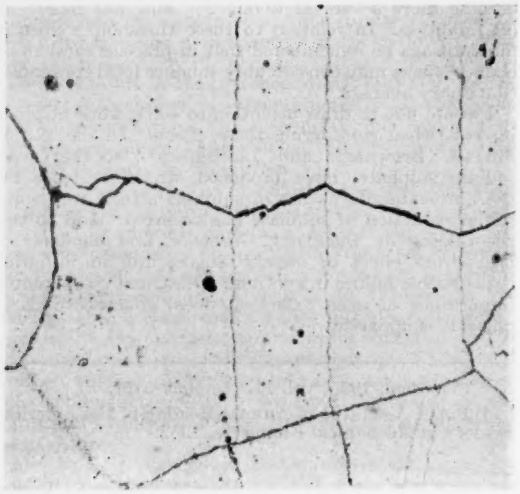


Fig. A—Transgranular stress-corrosion crack in stainless steel × 700

two classes of steels discussed on p. 129 is due to metallographic features, the nature of the surface oxide film or some other characteristic of the material.

Mr. **M. C. T. Bystram** (Murex Welding Processes Ltd.): The stress corrosion of stainless steels is a problem which is well worth studying since the mechanism of its initiation and propagation and the methods for preventing its occurrence may alter according to the prevailing conditions. I encountered a recent instance of stress corrosion in which corrosion occurred in a welded vessel made from 25–20 stainless steel welded with 25–20–Nb electrodes during service at temperatures of the order of 650° C in the presence of $ZrCl_2$. The stress corrosion cracking started intergranularly in the weld and propagated transgranularly in the direction of two spots about 3 in. away from the weld where the arc had been struck accidentally. It would seem that in this instance the mechanism of the process could not be associated with martensitic products in view of the stability of austenite in 25–20 steel and the high temperatures involved.

I would like some guidance on the best method of avoiding stress corrosion in welded stainless-steel vessels, and in particular, whether it is better to increase the corrosion resistance of an equipment by the choice of more highly alloyed steels or by the application of stress relief which, in my opinion, may in many instances produce more local stresses in a welded vessel than exist in it in the as-welded condition either because of rapid and uneven cooling from the stress relieving temperature or because of the precipitation of minor phases with slower rates of cooling. In this connection, it is possible to cite the example of weld metal of the 19–9–Nb type which without heat treatment, will exhibit a percentage elongation of the order of 45%, but which is likely to have an elongation of 29% or less after stress relieving for two hours at 800–900° C with a 24-h period of furnace cooling. This reduction in ductility is probably associated with precipitation of carbides and the formation of σ phase.

On the other hand, if, to avoid this effect, a heat treatment at 950° C is chosen with the vessel cooled quickly (in 2–3 h) through the temperature range 900–600° C by withdrawing it from the furnace, stresses may be induced because of uneven shrinkage effects. Therefore it seems to me that from the aspect of avoiding stress corrosion cracking in a welded vessel, the use of a suitable alloy is preferable to the application of stress relief after welding.

Dr. **C. Edeleanu** (Tube Investments Ltd., previously at the Brown Firth Research Laboratory, Sheffield): There is a great deal of agreement between the work done in Sheffield and that in the present papers. However, I think the following points require further discussion.

The authors seem to feel that the presence of δ-phase may be harmful, which is surprising. Figure *B*, for instance, shows the time to failure of a number of 18–8 Ti steels with different δ-content tested as tensile specimens at an applied load of 20 tons/in². Very similar results have also been obtained with sheet specimens tested as **U**-bends. If any conclusion can be drawn from these tests it is that the δ phase leads to an increase in resistance. The same conclusions were reached on plain Cr–Ni steels previously but, when it came to the Cr–Ni–Mo steels, the position became more complicated because in this case interconstituent stress corrosion cracking, Fig. *C*, was a common feature. In this case a high δ seldom led to much increase in resistance.

My second point is about the division of the steels into classes. I agree that some steels are more resistant than others but I am not sure that a division should be made in these particular cases except in so far as to say that the steels, as received, were in different conditions. A proof stress of 70,000 lb/in² as shown in one case, surely could not be due to coiling and uncoiling stresses. My feeling is that certain high-δ steels, as mentioned above, could be classed as more resistant and also some of the fully austenitic high nickel steels, but otherwise all the 18–8 steels are much the same.

On the theoretical side I still find it rather difficult to accept that ' chemistry ' can do damage at the rate at which cracks advance. I would like, if possible, to see the electrochemical argument put in greater detail when the present data, which I understand are being obtained, are available. If we are to believe that nothing much happens during the first 2–3 h of the specimen's life and that a crack then forms and causes failure rapidly, surely we should expect rather a small number of cracks on each specimen. This is so in many cases of stress corrosion but it is not so with the austenitic alloys. Some of the authors photographs show this to some extent but Fig. *D* shows it very clearly in a section through a specimen just under 0·2-in dia. Surely it is difficult to believe that all the cracking seen in Fig. *D*

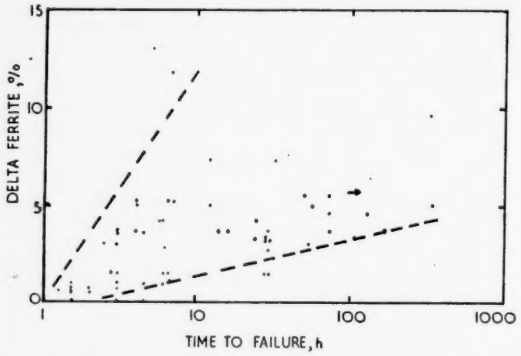


Fig. B—Effect of δ-control on the susceptibility of various commercial 18–8–Ti steels tested at 20 tons/in² in boiling 42% $MgCl_2$ solution

started in the last few minutes of this specimen's total life of 3 h.

My own feeling is that the potential measurements do not really indicate the cracking which occurs in the initial stages. I think I am correct in saying that proof stresses are also surprisingly little affected by a small amount of cracking. My impression is that cracking starts earlier than the authors have indicated and that as the effective load increases, because of the reduction in area, the cracking does become comparatively rapid. It is only then that a change is clearly seen on the potential/time curve. This would explain the great similarity between the end of the various curves obtained by the authors. The real criticism of the authors' theory is that it does not seem to help in showing why transgranular cracking occurs in only a relatively small number of alloys and why the Ni content of the steels is very important. Other effects such as that of cold work finds no logical explanation.

The quasi-martensite theory has its faults, but perhaps the faults are not so great if sufficient emphasis is given to the 'quasi' part of the expression. I feel that we do not really know what material is being formed when austenite is deformed above the so-called *m.d.* point, but I feel it is a material which deserves a name since it has the strange property of being easily etched preferentially. I feel reasonably sure that this is a field where the metal physicists could well give us some assistance if in no other way than to tell us that what I call quasi-martensite is the same thing as the highly strained material visualized by the authors.

Finally, something which may interest those concerned with the practical aspects of the problems, is that work to be published shortly in the *Journal* carried out in water and steam at high temperatures and pressures has shown that the phenomena is exactly the same as that in concentrated chloride solutions. Information such as that in the present papers is very pertinent even as far as the normal service failures which generally occur in water and steam service.

Fig. C—Interconstituent stress corrosion sometimes found in 18–8 Mo steels tested in boiling 42% $MgCl_2$ solution

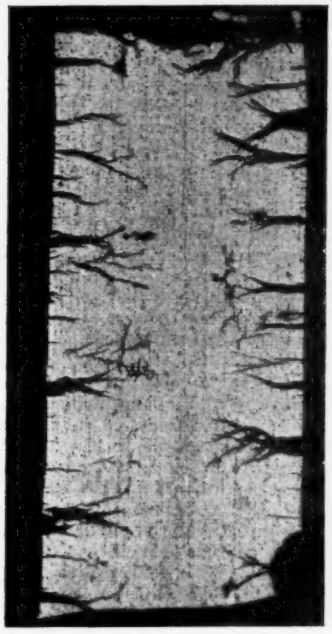


Fig. D—Typical central section through stress–corrosion specimen, tested at 20 tons/in² in boiling 42% $MgCl_2$ solution × 8

AUTHORS' REPLIES

Dr. Hoar: I agree with Dr. Parkins that stress-corrosion cracking is becoming a more serious problem in practice and that work on the practical cases that now occur is needed. It is one of the few corrosion problems that we are beginning to tackle in time and not 20 years late.

In his implied criticism of our use of small specimens, Dr. Parkins suggests that we have studied mechanical rupture triggered off by corrosion pitting. This is quite unacceptable. The cracks in all our specimens are typical transgranular stress-corrosion cracks, as illustrated in Parts I and II of our paper, and any corrosion pits are of depth small compared with the thickness of the specimens. Of course, with much thicker specimens the time of cracking may be made large compared with the induction period. It can also be made large compared with the induction period by almost eliminating this by removing the initial oxide film; in such cases, as reported in the first paper, cracking took place for substantially the whole time the specimen was exposed, fracture occurring in a few minutes. Thick specimens under such conditions would crack continuously for hours before fracture; Fig. *D* of Dr. Edeleanu's contribution, which incidentally shows cracking qualitatively identical with that of our specimens, on a ten-times larger scale, is probably an illustration of this. As to the relationship between applied stress and time to fracture (*not* cracking time, as misquoted by Dr. Parkins), our results are in harmony with, for example, those obtained by Franks, Binder, and Brown using much larger specimens of an entirely different type.

We do not 'claim' that stress, internal or applied, is not important in the induction period for our case of stress-corrosion cracking in 18–8 steels: we state it as experimental fact. The different case of mild steel is

quite irrelevant, although we entirely accept Mr. Parkins' interesting experimental demonstration of the different facts there found.

We have nowhere inferred that there is no ' metallographic aspect ' to transgranular stress-corrosion cracking in 18–8 steels: on the contrary, we believe that all the evidence shows that the crack progress, although in general perpendicular to the main applied stress, is modified by tesselated stresses, change of structure, and indeed all the microscopic heterogeneities that can be termed metallographic. Dr. Parkins' Fig. *A* is of interest, if a little difficult to interpret fully, and certainly adds to the general evidence for the modifying influence of structure. However, with regard to the differences between our two suggested classes of steel on p. 129 of our paper, we think that there is at present insufficient evidence to assign such differences to one cause only, although susceptibility to cold work (with its structure-modifying effects) is probably important.

Dr. Edeleanu's results on the favourable influence of δ-ferrite are of much interest. In our experiments with a limited number of steels, we obtained no evidence ' for or against ' δ-ferrite, which is not surprising in view of the wide scatter shown in Dr. Edeleanu's Fig. *B*. We agree entirely with Dr. Edeleanu that all 18–8 steels will stress-corrosion-crack. Variations from cast to cast are probably as important as deliberate variations of composition, and variations of mechanical and heat treatment are also important. Nevertheless, since all 18–8 steels stress-corrosion-crack in many hot and cold chloride solutions, we think there is a general mechanism involved; and indeed Dr. Edeleanu's own recent work on stress-corrosion cracking in steam, where he finds that chloride contamination and liquid condensation are necessary to give the cracking, is in harmony with the general electrochemical theory that we have outlined in the papers.

Since these papers were written, we have greatly developed the electrochemical theory. For example, by determining the electrode-potential/current-density relationship for oxygen reduction on the specimen surface, we are enabled to estimate the current available to flow into the cracks with fair accuracy, from potential measurements. It transpires that there is plenty of available current to account for a purely ' chemical ' or ' anodic dissolution ' advance of the cracks at rates up to 1–2 mm/h. Furthermore, cathodic protection experiments have shown that cracking can be slowed down or stopped entirely by the application of an external e.m.f., and other observations have led to an explanation of the influence of nickel content on the anodic reaction. Since these results will very shortly be published in detail, perhaps we may refer Dr. Edeleanu to them in reply to his doubts about ' chemistry ' as the means of crack propagation.

Dr. Edeleanu's most interesting criticism is his remark that a long induction period followed by a short cracking period might be expected to be associated with a small number of cracks. We doubt this as a general deduction, because we see no reason why many cracks should not be initiated at approximately the same time. The particular specimen (Fig. *D*) with which Dr. Edeleanu supports his point seems to us rather to refute it. This specimen, ten times the diameter of ours, shows many long cracks, up to 2 mm, and many short ones. Its time to fracture was 3 h; and a crack-propagation rate of $\frac{2}{3}$ mm/h is entirely in harmony with our observed rates. We think therefore that this particular specimen had a short induction period (perhaps the magnesium chloride was acid?) and a relatively long crack-propagation period, during which further cracks were initiated. Alternatively, the short cracks may have been initiated early on but prevented from developing because the larger ones ' getting ahead ' relieved the stress at the advancing edges of the smaller ones, and also tended to take more than their ' fair share ' of the total available anodic current.

With regard to quasi-martensite, Dr. Edeleanu suggests that provided we call it as ' quasi ' we are all in agreement, since our ' highly disarrayed metal ' may well be the same thing. Certainly we believe that *any* easily dissolving material may well be important, especially if the yielding just ahead of the advancing crack can produce it. But we think all the evidence is against any pre-formed ' susceptible paths ' in the present case of stress-corrosion cracking, and we think that evidence such as Mr. Bystram's example shows that martensite-like material is not *necessary* for the phenomenon to occur.

In answer to Mr. Bystram, whose comments are very interesting, we can give little firm guidance as to how stress-corrosion cracking should be avoided. However, stress-corrosion cracks are not initiated on 18–8 steels in hot chloride solutions if the metal is mildly polarized cathodically, to an extent corresponding to a current density of some 50 $\mu A/cm^2$. Also, if a crack has started, a slightly higher polarization completely stops crack propagation. There may be practical value in cathodic protection in certain chemical plant; if a particular area is known to be susceptible to stress-corrosion cracking, there seems no fundamental difficulty in supplying internal local cathodic protection. Practical experiments on such lines are very desirable. However, corrosion inhibitors, the removal of oxygen and adjustments of alloy composition (with the possible exception of increased nickel) do not seem very hopeful expedients in the light of our present knowledge.

May I conclude by again emphasizing that these papers deal with transgranular stress-corrosion cracking in 18–8 steels. That many of the phenomena are quite different from those of the intergranular cracking of mild steel, brass, and light alloys is not in the least surprising. We do not suggest for a moment that the mechanisms of transgranular and intergranular cracking are the same.

Dr. **Hines:** Dr. Hoar's reply to the main discussion covers the ground admirably, but the point concerning the effects of the δ-phase does, I believe, merit further consideration. From Dr. Edeleanu's contribution it seems that there is no real disagreement between us on this matter, because the results he quotes in Fig. *B* and our results are probably not comparable. We examined three steels containing δ: 18–8–3 Mo(*a*), 18–8–3 Mo(*b*), and 18–10–Mo–Ti, all of which contained appreciable amounts of Mo, while Dr. Edeleanu found that δ-phase had an effect on the resistance of plain Cr–Ni steels and of 18–8–Ti steels to stress-corrosion, but little or no effect on the Mo-bearing steels. Thus if the Mo-bearing steels are considered as a separate class, our results do not contradict Dr. Edeleanu's.

With regard to the 18–8–Ti steels, examination of numerous service failures does provide some support for the view that δ-phase is desirable. Failure in steel containing appreciable amounts of δ-phase appear to be much rarer than would be expected if such material were as susceptible to stress-corrosion cracking as fully austenitic material. In one case the effect was particularly striking; extensive cracking had occurred in and to one side of a weld between two piece of plate, and on examination it was found that the cracked region was almost fully austenitic while the second plate contained some 5% ferrite. In this case one might suppose that the conditions on either side of the weld would be reasonably similar. Thus, while the beneficial effect of δ cannot be regarded as proven, it is at least a pointer in a field where

indications of a means of improvement are few.

CORRESPONDENCE

Dr. **H. K. Farmery** (Fulmer Research Institute) wrote: The mechanism suggested by the authors for crack propagation in austenitic stainless steels, namely the continuous anodic dissolution of highly stressed and strained metal, is extremely interesting. There are cases in other systems, however, where very little actual corrosion is needed to produce fracture in a metal or alloy which would not occur in the complete absence of corrosive influence. With the Al–7%Mg alloy for instance, the rate of crack propagation is affected by the particular heat treatment to which the material has been subjected and in most cases cracking is apparently much too rapid to be explained by a gradual eating away of metal at the tip of the advancing crack. Have the authors found any such variation with the different types of austenitic stainless steels?

AUTHORS' WRITTEN REPLY

The **Authors** wrote: Dr. Farmery rightly points out that the rate of (intergranular) crack propagation in the stress-corrosion cracking of aluminium–magnesium alloys is markedly influenced by heat-treatment, and can be so rapid as to preclude a mainly electrochemical dissolution mechanism in that case. With different compositions and heat treatments in the 18–8 stainless steels, however, the variations in the rate of the transgranular crack propagation are very much less, and the absolute rates are in harmony with the electrochemical mechanism we have proposed.

CORROSION RESISTANCE OF 18-8–Ti Cr–Ni-STEELS

This discussion was based on the paper by Messrs. E. J. Heeley and A. T. Little entitled "Corrosion Resistance of Some Austenitic Cr–Ni Steels of 18/8/Ti Composition" which appeared in the March, 1956, issue of the *Journal* (vol. 182, pp. 241–255). Mr. **A. T. Little** (I.C.I. Ltd., Dyestuffs Division) presented the paper.

Dr. **T. P. Hoar** (Department of Metallurgy, University of Cambridge): Here is a case of corrosion where a material, 18–8–Ti steel, is being used for service possibly more onerous than that for which it was originally designed. Under certain conditions it is clear that 18–8–Ti steel is by no means uncorrodible. Users will always try to use a metal or alloy for more onerous service than the makers originally intend, and its failure is then no reflection at all upon the manufacturer. On the other hand, when such particular cases of failure are investigated and understood they often lead to new advances.

In the 18–8 group of steels, and perhaps *a fortiori* in the even higher-alloy Cr–Ni steels, it is generally believed that the corrosion resistance is due to the formation of a very stable and self-healing film on the surface of these materials. This, however, is possibly the most important, although not the only reason for their stability against many corrosive environments. In recent work on the stress-corrosion cracking of this class of stainless steel, we have found that nickel seems to have some intrinsic effect in raising the potential of iron, consequently ennobling the alloy. It seems possible that this effect may become of some importance in cases, such as the present, where film-breakdown seems to have occurred through local chromium impoverishment.

Titanium stabilization is clearly not perfect. Have the authors any evidence of a similar 'sensitization' effect in Nb-stabilized steel? Again, they have shown already that with carbon below 0·06% the phenomenon is reduced; how about the effect of extra low carbon, i.e. 0·03% maximum? Although extra low carbon steels do not require stabilization with either Ti or Nb to resist ordinary weld decay, can we expect them to be immune from a '650° C effect'? Equally, do the effects produced by the 1300° C heat treatment occur in extra low carbon steels or in Nb-stabilized steels, and is the improvement in fairly low carbon (0·06%) steels also found with Nb as stabilizer?

Finally, have the authors come across this kind of sensitization to attack by media other than nitric acid?

Dr. **C. Edeleanu** (Tube Investments Ltd.): The authors have gone a very long way in showing it is possible to use the 18–8–Ti type of steel under very severe nitric acid conditions and much credit must be given to them for this. However, it is only fair to make it clear that for this particular application the 18–8–Ti type of steel is only a second best and that anyone else who might be considering building plant for nitric acid or other highly oxidizing solutions would be well advised to think about Nb-stabilized steels or even very low-carbon unstabilized steels. In certain cases the 15–17% plain Cr steels may also be found useful, provided proper post-welding treatments are given.

The 18–8–Ti steel is no 'better' or 'worse' than the equivalent Nb-stabilized steel any more than, say, an apple is better or worse than a pear. Once this fact is recognized any suggested alteration in composition must be considered with respect to the effect it will have over all its properties. Therefore I think it is fundamentally wrong to use nitric acid as a sorting out 'test solution' for a steel which is not made primarily for use with this acid. The authors' suggestions about the direction in which the composition of the 18–8–Ti steels ought to be allowed to drift may improve their nitric acid resistance but it has not been shown that this drift, if allowed to proceed too far, will improve other properties.

There is little mention in the paper, for instance, about the effect of the composition on the δ-content of the steels; do the authors not consider it very important to keep good control of the δ-content?

A second point which may require some thought is that as the carbon is lowered there will be a temptation to drop the titanium as well. If this is done any slight carbon pick up which may occur during manufacture or fabrication may have a relatively greater effect than at the higher titanium levels now used.

Mr. **M. C. T. Bystram:** The conclusions from this paper confirm the general advantages of carefully balanced alloying in the stainless class of steels. From a practical point of view it has been established in this work, that welded 18–8–Ti steels, with carbon of the low order of 0·6% and adequate stabilization with titanium will be free from 'knife edge' attack in the zone immediately adjacent to the weld, where the steel was reheated above 1300° C. For this reason the slightly more expensive Nb-stabilized grades are normally preferred for media containing nitric acid.

We can shortly dismiss from this discusssion the medium temperature impairment of stainless steels, known as 'intergranular corrosion' which occurs mainly in the range 600–750° C, as there appears to be no difficulty in avoiding this trouble with adequately Ti-stabilized grades even with the higher carbon content (0·1%), especially if the balance of ferritizers and austenizers in the steel results in reasonably low ferrite contents, which eliminates the possible negative effects of σ-phase formation.

The knife-edge attack of high-temperature corrosion impairment is of a greater interest because there appears to be no explanation of its exact mechanism and the same area adjacent to the weld can also suffer from other defects like hot cracking, which also happens at room temperatures above 1300° C and which is associated with segregation tendencies.

It would seem in fact that while the temperature range of say 1100° ± 100° C might be associated with homogenization, the range between 1250° C and solidus appears to be causing dehomogenization (segregation).

I had frequent opportunities of confirming that the stainless material immediately adjacent to the weld has a structure intermediate between plate and weld, in that the feather-shaped ferrite islands (as in weld metal) are arranged along the direction of ferrite stringers in the plate.

Above 1350° C δ-ferrite appears in a considerable amount (40% and more) and there is a tendency to enrichment of ferritizers in ferrite and austenizers in austenite, hence it seems likely that the mechanism contributes to concentration of the very strong ferritizer titanium in δ-ferrite, and of carbon in austenite, thus preventing their recombination without a homogenizing treatment at 1100° C. The fact that ferrite forms a diffusion barrier for carbon, prevents the recombination of Ti and C at lower temperatures (e.g. 750° C). I hope that this hypothesis may contribute to the clarification of the experimental evidence contained in this paper.

The generally better performance of Nb-stabilized steels in nitric acid could be explained by greater stability of Nb carbides at high temperatures and weaker ferritizing effects of columbium. In line with these arguments it would be better from the experimental point of view to use samples with 'isoferritic compositions.' In other words, compositions with Cr and Ni balanced to give approximately similar ferrite contents in the coatings.

To prove the dehomogenizing tendencies in the range above 1250° C it would be interesting to study the resistance to high-temperature inpairment of Ti-stabilized (argon-arc) weld metal with carbon above 0·06% as compared to plate of similar composition. The bulk of the weld would be solidifying comparatively quickly from more homogenous liquid and could differ from the sluggish region in the plate adjacent to the weld.

Mr. **H. T. Shirley** (Brown-Firth Research Laboratories): It seems desirable to clarify some aspects of the position of the titanium-stabilized 18–8 steels.

Over the last 25 years many hundreds of tons of such steels, to ordinary specifications, have been successfully used under a wide variety of conditions involving nitric acid. During this period considerable developments have occurred in chemical processes and more recently the severity of conditions has in a number of cases made it necessary to consider in detail the response of the various special steels under conditions peculiar to strong nitric acid, particularly at higher temperatures.

Where such conditions have to be met, improvement in resistance can be obtained by adjustment of the composition balance in Ti-stabilized materials or by changing to Nb-stabilization. The authors have described their work under the first alternative and have given most useful information defining C–Ti ranges for safer working under the difficult service conditions discussed. In the laboratory, results of our own tests have broadly agreed with their findings, and although the suggested modification in composition ranges involves more stringent specifications than hitherto, these will be met where necessary.

The alternative is to use Nb-stabilization. Some hesitancy has arisen from the fear that difficulty might be experienced from the more strategic position of Nb compared with Ti, but there is no indication of any serious difficulty of this type at the moment. In many cases this would seem the simpler and safer alternative.

On the matter of mechanisms, it is probably unnecessary to complicate the general picture for nitric acid beyond the ordinary time/temperature effects associated with carbide precipitation in relation to intercrystalline corrosion. For example, the knife-edge attack at welds can be explained in terms of the very high local-temperature treatment during welding making this region much more susceptible to subsequent precipitation effects. It must be remembered here that hot strong nitric acid is a most powerful reagent for showing up differences in effective chromium content. At the same time, under some conditions there does appear to be an effect from titanium not wholly explained by incomplete carbide stabilization. Thus Stewart* showed that an unstabilized steel could behave better than a Ti-stabilized one in boiling 65% nitric acid. Where such effects occur they emphasize the greater safety from Nb-stabilization.

AUTHORS' WRITTEN REPLIES

The **Authors** wrote: Dr. Hoar's remarks appear to suggest that the 'nitric acid' corrosive conditions to which 18–8–Ti steels were being exposed in the chemical industry are excessively onerous for this type of material, but as pointed out by Mr. Shirley, large tonnages of this steel have been successfully used over a wide range of such conditions.

In fact, the 18–8–Ti material which we employ in itself possesses a satisfactory resistance to the corrosive conditions under consideration but unfortunately is susceptible to impairment of corrosion resistance when subjected to certain thermal treatments encountered in fabrication processes. This susceptibility to impairment of corrosion resistance can, of course, be reduced by adopting the chemical composition given in the paper.

The Nb-bearing austenitic steels, to which reference has been made by Dr. Hoar and the other contributors to the discussion, are outside the scope of a paper on the Ti-bearing austenitic steel but it is perhaps desirable to point out that the Nb-bearing steel should not be regarded as being entirely immune from susceptibility to impairment of corrosion resistance when subjected to the thermal treatments applied in the course of fabrication processes.

We have found, for example, that the 18–8–Nb type of steel of usual commercial production with C content about 0·10% exhibits the '600–750° C effect' if the Nb content is around the lower level usually specified for this element i.e. 8 × C%. Whilst the impairment is less in magnitude than in the corresponding Ti-bearing steels, the susceptibility may be eliminated by raising the Nb content to about 12 × C%. Although the Nb-bearing steels are much less susceptible to the '1300° C effect,' plant experience has shown that this problem can be serious when very onerous 'nitric acid' conditions

* R. S. Stewart: *Metal Progress*, 1947, vol. 52, pp. 971–973.

prevail. No data are available regarding the effect of lowering the C content on the susceptibility of the Nb-bearing steels to the two forms of corrosion impairment, but it would seem reasonable to assume that such a step would be beneficial, and reduce the amount of expensive Nb addition required.

Regarding the 'unstabilized' 18–8 steels of very low C content i.e. 0·03% maximum, the small amount of work which we have carried out on such steels has indicated that they are superior to either of the stabilized steels as regards susceptibility to the ' 1300° C effect ' and the ' 600–750° C effect '.

We have had no knowledge of either of these two forms of impairment having occurred in any other of the corrosive media for which the stabilized 18-8 type of steel is usually employed.

Dr. Edeleanu refers to a possible temptation to drop the Ti content in conjunction with lowering the C content in the Ti-bearing steels. We consider that it is necessary, when the C is restricted to 0·06%, to have considerably less Ti than is found in the more usual 18–8 Ti steel containing about 0·10% C, and we suggest that with the former C content, the Ti should be held within the range 0·35–0·55% to obtain the optimum corrosion resistant properties of the lower C steel.

We are well aware, of course, of the hazards which could result from excessive C pick-up during fabrication of the stabilized 18–8 steels and appreciate that the effect on the corrosion resistance would be more serious in steels of lower C content. With the employment of correct fabrication techniques and procedures appropriate to the material being handled, however, we doubt whether the extent of C pick-up would significantly affect the corrosion resistance of the 0·06% C max. 18–8 Ti material.

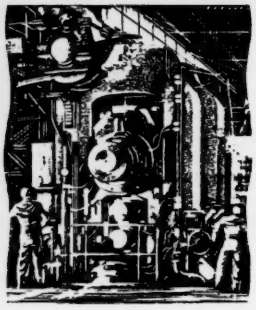
In this connection, we would point out that C pick-up during fabrication could be disastrous in the case of the very low-C unstabilized steel, the employment of which Dr. Edeleanu suggests is preferable to the 18–8–Ti type steel for handling nitric acid or other oxidizing media.

The 18–8–Ti steels normally contain small quantities of δ-ferrite phase, the amount depending *inter alia*, on the balance of alloying elements present, and in the 0·06% C max. steels which we have examined, the amount of δ-ferrite is, in general, smaller than in the corresponding steels containing 0·10% C. We do not consider, however' that the variations in δ-ferrite content, over the range normally encountered in the Ti-bearing steels, have any significant effect in respect of susceptibility to ' 600–750° C' impairment of corrosion resistance. The amount of δ-ferrite formed at high temperatures (*circa* 1300° C and above) may, however, have some bearing on the ' fissure ' wastage phenomenon, but we have not investigated this aspect of the problem in detail.

Mr. Bystram suggests that formation of σ-phase (from δ-ferrite) may be responsible for the ' 600–750° C effect ' found in the 18–8–Ti steels containing about 0·10% C and that this impairment may be avoided by employing a steel composition ' balanced ' in respect of ferritizers and austenizers. In reply we would say that in our investigations we have tested a number of wrought 18–8–Ti steels containing about 0·10% C with varying amounts of δ-ferrite and also with practically no δ-ferrite, and there has been no indication that the δ-phase in any way influences the degree of the corrosion impairment resulting from heating those steels in the temperature range 600–750° C.

We are most interested in the mechanism suggested for the ' 1300° effect ', viz. segregation of the ferritizers and austenizers, and can confirm that in many of the 18–8–Ti–steels heat treatment in excess of about 1300° C results in the formation of considerable quantities of δ-ferrite.

Mr. Shirley refers to the mechanism governing the fissure wastage (' 1300° C effect ') which may occur at welded joints in 18–8–Ti material of certain compositions. We agree that the true explanation of this effect embodies something more than increased susceptibility to chromium carbide precipitation in the narrow, heat-affected zone adjacent to the weld. In fact, we have shown by laboratory tests that an unstabilized steel containing 0·09% C is not susceptible to impairment of corrosion resistance on heat-treatment at 1300° C. It would appear, therefore, that the presence of a stabilizing element (e.g. Ti or Nb) is necessary for the austenitic steel to develop susceptibility to the ' 1300° C effect.'



IRON AND STEELWORKS ENGINEERING

The Iron and Steel Engineers Group

IRON AND STEEL ENGINEERS GROUP

REPORT OF THE THIRTY-FIRST MEETING

THE THIRTY-FIRST MEETING OF THE IRON AND STEEL ENGINEERS GROUP of The Iron and Steel Institute was held in South Wales on Thursday and Friday, 4th and 5th October, 1956. The technical sessions were held in the Grand Pavilion, Porthcawl, with Mr. W. M. LARKE, Chairman of the Group, in the Chair.

At the second technical session, on Friday, a paper on " Planning the Conversion of a High-Lift Slabbing Mill to a Universal Mill " was presented by Mr. C. E. H. Morris and Mr. R. N. Dale (Steel Company of Wales, Ltd.) and was discussed. The paper was published in the April, 1957, issue of the *Journal* (pp. 532–552). A report of the discussion is given below.

THE CONVERSION OF A SLABBING MILL TO A UNIVERSAL MILL

Mr. **F. E. Probyn** (Richard Thomas and Baldwins Ltd., Ebbw Vale): I should like to know the minimum time that has been allowed for the grout to harden under the roller-table side frames in the manipulator section. From our experience this is an important point.

Mr. **R. N. Dale:** Grouting to table frames can be done fairly early in the programme. There will still be about six days before it is necessary to work the table.

Mr. **R. A. Lake** (Richard Thomas and Baldwins Ltd., Redbourn): I failed to understand from the paper how the authors intend to modify their manipulator carrier arrangement, which is a sliding shoe. In talking to the mill engineer a few months ago I gathered that, although not very troublesome, it requires more attention than our simple trod-wheel arrangement on an elderly cogging mill, on which we have converted the trod-wheels to taper roller bearings.

The limitation of individually driven rollers, which are otherwise more desirable than a group drive arrangement, is the torque available at any one roller to move the piece, when the latter is in contact with only one or two rollers of the roller table, a condition which occurs frequently. Would it not be better to couple the two feed rollers together so that the torque equivalent of 70 h.p. would be available, even though there was contact with only one roller?

Mr. **C. E. H. Morris:** In the existing roller-table framework on the entry side of the mill, there are spreader beams on which were machined sliding surfaces. With the wear which has taken place during the last few years, I should not describe them at the moment as machined. The underside of the manipulator head itself rests on two shoes, and these shoes slide backwards and forwards on the beams.

In the designs which we are now proposing, the fixed slide will be outside the table frames, and not between the table frames, with the shoes mounted on the slide, only this time upside down. These shoes will be fixed and the manipulator itself will be fitted with a machined sliding surface underneath, which will now slide on the fixed inverted shoes, instead of the shoes sliding on the fixed beam.

We must admit that the troubles which we have met in the past with individually driven feed rollers have been partly due to the fact that the rollers were driven together through a gear box, and are of different diameters and on different pass lines. We now consider it better to install 35 h.p. slow-speed motors, but these motors are actually 200 h.p. frame size operating at low speed. Generally speaking, we should prefer, if possible, to avoid use of a gear box on a drive so close to the mill, where the ingot emerges and does not necessarily contact each roller so that the load is unequally distributed. If they are individually driven the design is more efficient and neater.

Commander **J. I. T. Green** (B.I.S.R.A.): Regarding manipulators; the operators will have to keep the slabs central at the back of the mill, and I do not understand how they can see to do that. Should not the manipulators be designed as automatic centring devices?

Mr. **Morris:** All manipulators at the back side of mills are partly centring devices. Our hope is that they will not be used as slab stoppers and brought together to stop the slab as close as possible on the outgoing side of the mill. They will be used to centre the slab when it is on the roll table, but in addition we intend to use a television camera on the outgoing side of the mill. We have used television in a slabbing mill on one occasion when both the finger-tilt arrangement broke down on the entry side and we lost the main manipulators on the

exit side. Under these conditions we have used a large hook suspended by a crane, which tilts the slab or ingot, but when this has to be done on the outgoing side it becomes a little more difficult. Our operators soon became very adept at using a television screen to observe the operation of the crane hook on the outgoing side of the mill; it will be a permanent installation, and the operators can confirm that the slab is central for its return pass.

Mr. **E. R. Henschker** (Loewy Engineering Co., Ltd.): I noticed a reference to the use of the porter bar from the finishing mill in connection with taking out the breast roller. Could we have some information on what kind of bearings were used on the first breast roller and whether the breast roller had to be changed frequently?

Mr. **Dale:** The bearings are phosphor-bronze and they last for at least a year. The rollers last 1–5 years, so that the operation of changing rollers or bearings is most infrequent. It is quite practicable to use the porter bar from the finishing mill with a special adaptor.

Mr. **P. E. Peck** (British Thomson-Houston Co., Ltd.): The authors mentioned the length of table available between the 'weigh and turn' and the mill as a limitation on rolling. The position chosen for the edging mill will further reduce that length. What factors dictated this? There appears to be much more room on the outgoing side of the mill.

Mr. **Morris:** It is not so much the physical limitation of the ingot turning position. It is rather that we are bringing two ingots up to the mill ready to roll while we are still finishing the ingots in the mill. One of the ingots which is ready for rolling, having been weighed and turned, has to be held on the mill side of the ingot-weighing platform, and on the comparatively rare occassions when the slabs run out so long that they would reach this distance there is a limitation.

On the entry side of the roller tables, however, we are putting an extension of about six rollers. This is partly because we can then have two ingots ready to go into the mill and have them clear of the ingot-weighing platform, so that the full physical distance of the mill is clear for running out tandem-rolling slabs.

Apart from that, the position of the vertical rolls on the entry side was dictated in part by prejudice. We believe that the finger-tilting and vertical rollers and operating pulpit should all be on the same side. We even considered that we might have to move the operating pulpit to the other side of the mill, but that would be extremely expensive; it would have to be an additional installation. Having decided to leave the pulpit where it was it naturally followed that we should also leave the vertical rollers on the entry side.

As far as the motor room itself is concerned, it is slightly easier to build the motor base on the entry side rather than the outgoing side.

Dr. **L. N. Bramley** (B.I.S.R.A.): We have not mentioned 'automation' so far. I should like to ask whether the future possibilities of the mill have been fully discussed and whether all the auxiliary drives are being planned so that if it becomes necessary to install closed-loop controls on these drives it will be possible to do so. It is very important that the drives should be Ward–Leonard control and not contactor control, since the implementation of remote-position control on contactor-controlled drives presents much greater difficulty.

I should like to ask whether the screwdown dial, at present situated inside the pulpit of the mill, is to be left there in the new set-up. To my mind, it is wrong there from an ergonomic point of view because operators alternately watch the screwdown dial and then the ingot and there is a great deal of focusing and re-focusing of the eyes. The possibilities of projection television might be examined with the object of eliminating this constant re-focusing.

Mr. **Morris:** The answer to the first question is that the manipulators and screwdown will all be Ward–Leonard control. I would refer the second question, about the screwdown dial, back to Dr. Bramley, because the success of the position control systems, one of which we are about to try out, will determine to a great extent any new lay-out we adopt in the operating pulpit.

Mr. **W. N. Jenkins** (B.I.S.R.A.): Is consideration to be given in the new mill to the application of translator control to screws and manipulators?

Dr. **Bramley:** If you do consider it and use it you will not need a screwdown dial although it will probably be advisable to retain it for the present.

Mr. **W. Bailey** (Briton Ferry): The use of models has a great many advantages. One of them is apparent when you have directors who are not technical and who cannot read drawings clearly. It is easier to understand the situation when you get down to models, which are three-dimensional.

The relation between the time in the pass and the manipulation time has never been properly established and the Steel Company have had to do this work. Some of the estimates on blooming mills have suggested that the figures are 50% in the pass and 50% in manipulation, but that is very empirical. Since this not only establishes phenomena for motor design but also gives engineers data for improving mill design, I wonder whether Mr. Morris could give some rough figures.

Mr. **Morris:** In supporting what Mr. Bailey said about the use of models, I should tell you that our models were made by our apprentices under the supervision of the foreman-carpenter. Davy–United made the model of the new mill. We shall shortly return it to them and it will come back to us as a completely working model, with gears cut and everything rotating as required.

All the information on relative times in and out of the rolls has been published by B.I.S.R.A. and by Mr. H. G. Jones, our operational research manager, who was responsible for the supervision of all the tests which were made, in conjunction with the rolling mill management. Mr. Jones perhaps would like to say a few words on the subject.

Mr. **H. G. Jones** (The Steel Company of Wales Ltd.): We have made numerous exact measurements on the time breakdown during the slabbing process and the results have been published in the August issue of the *Journal*.* The proportion of time in contact with the rolls will vary with the ingot and slab being rolled, but in one particular case it was 35%.

Mr. **K. Bose** (Indien Gemeinschaft): I should like to know for what maximum size of ingot the mill is designed when rolling down to the minimum slab section, the number of horizontal and vertical passes, and the approximate time taken. Does the author think that automatic pre-set screwdown control would increase production and lessen human effort? Will this mill have this control?

The authors spoke about the restricted headroom on the existing mill building in installing this overhung edger. Did they consider the use of the underhung type of edger?

Mr. **Morris:** The maximum weight of slab which we have produced to date is about 30,000 lb.

Mr. **W. F. Cartwright** (The Steel Company of Wales): It is 37,000 lb.

Mr. **Morris:** That is from an 18-ton ingot, and is at

* H. G. Jones, D. T. Steer, and P. D. Dickerson: *J. Iron Steel Inst.*, 1956, vol. 183, pp. 375–386.

the moment about the maximum. On the present mill we can only produce a slab 68-in. wide, but we intend to increase this so that we can roll slabs 72-in. wide, which will eliminate any cross rolling necessary in the strip mill and result in an increase in tonnage.

The reference to pre-set screwdown control may not have been clear. During the last year or two B.I.S.R.A. have developed three main types of pre-set position control, one of which we are installing in the Abbey works mill. Installation will be complete in the next week or 10 days. The results of that will determine to what extent we adopt it in the future.

By the underhung type of edger I presume you mean an edger where the drive is contained beneath the mill.

Mr. **Bose:** Yes.

Mr. **Morris:** We felt that if it were possible to get a design in which the drive gear was above the mill and therefore free from falling scale and slightly more shielded from direct heat, it would be desirable to do so. That design we have now developed, and we believe that in having it overhead we have the better of the two choices.

Mr. **J. S. Caswell** (University of Wales): I would like to ask the authors what is the extent of the alterations to the foundations, because the foundations of the existing mill are of concrete which reached maturity long before the present mill was laid down. Have the authors any fears in this matter, because the time within which the work is to be completed is very short and perhaps not long enough for any new concrete to reach adequate maturity?

Mr. **Dale:** The new concrete work necessary can be done in the preliminary stages before the changeover. The extension to the existing roll-change pit will be done beforehand. The only concrete work necessary during the actual change over will be to cut out mature concrete and expose other existing mature concrete. We shall therefore not be troubled with the problem of deciding how long the concrete should have to mature.

Mr. **W. G. Beynon** (Admiralty): I wondered whether the authors did not think the pre-heats rather high and might result in an ultimate pre-stress of 15 tons/in^2.

Mr. **Dale:** It will not be as high as that because some of the pre-heat is necessary in order to pull up the allowances which are necessary to set the pieces in place. We have not decided how much pre-stress we shall put into them because we do not know how accurately we can make the fits with the spreader beams, but we shall probably allow so much temperature rise to pull it up solid and so much to pull it up to a stress possibly of 5–6 tons/in^2.

Mr. **S. Cornforth** (Dorman Long (Steel) Ltd.): I noticed that the drive to the vertical rolls includes a rather slender drive shaft 32 ft long with a heavy armature at one end and a set of heavy spiral bevel gears and spur gears driving the vertical rolls at the other end. When the piece leaves the rolls the stored energy in this shaft will be converted into kinetic energy in the masses at the extremities; and perhaps later when a certain amount of wear has taken place in the gears, couplings, and bearings, it will tend to produce an over-run of the vertical rolls. Has any provision been made for a torsional vibration damper on this drive?

Mr. **Dale:** The couplings of the shaft will be Holsett torsionally resilient couplings. Apart from that, the diameter of the shaft, which is 20 in., is sufficient to limit the stress to about 3 tons/in^2 for normal loading, due to the combined weight of the shaft and the torque of the motor. This, of course, would go up to approximately 9 tons/in^2 if we ever reached our peak horsepower, but that is not considered to be excessive. The Holsett couplings will take out a great deal of shock. We have not used them on anything as large as this, but the makers have put them on a main drive shaft for John Summers in their slabbing mill, which has proved successful.

Mr. **J. D. Dunlop** (Armstrong Whitworth (Metal Industries) Ltd.): Can Mr. Morris give any idea of the amount of time taken in planning this operation to date?

Mr. **Morris:** The development scheme reached the stage of estimating and initial design towards the end of last year (1955). It was at the beginning of this year (1956) that we started on the design of the mill which we were describing this morning and it has been proceeding steadily ever since then. We had to produce an approximate programme of the dismantling and re-erection time as long ago as November 1955, and since then all our attention has been devoted to ensuring that the design would meet the fairly tight programme that had been set. As recently as the middle of September we had the whole dismantling and re-erection programme checked by an independent engineer. Fortunately, he agreed within a few hours to the programme on which we had been working. The momentum has been increasing rapidly all this year and will reach its peak at the point of no return, at which time we can do no more alterations to the design and the work must be carried out.

Mr. **W. F. Cartwright** (Steel Company of Wales Ltd.): Some people may wonder why the 15-ton bottle-top tinplate ingot has to be of such a shape that it requires 14 in. of edging work. The reasons are as follows.

In operating strip mills it is desirable to have the largest possible coil, and our coilers were designed for 500 lb per inch of width, which means 15,000 lb for a 30-in. tin-plate coil. If a single slab came out of an ingot to make this coil the ingot would be 7½ tons, but 7½-ton ingots are small for use with 240-ton open hearths, and very small indeed when it comes to carrying from our further pits. This was realized in the very early days of 1947 and it was decided that the only way of avoiding these two difficulties of excessive numbers of ingots on each ingot train and too frequent journeys by the ingot buggy, was to have two slabs from each ingot, making a 15-ton ingot. Ordinary rimming steel cannot be made successfully higher than 72-in., and starting with a fixed height and a fixed weight the cross section is also fixed.

If the edging work was reduced then the thickness became excessive and the ingot was more or less square, and the only possible compromise was to have an ingot which was too wide and too thick, which for a 32-in. wide slab meant an ingot 50 in. wide and 30 in. thick. The excessive slabbing work which such an ingot involved resulted in the suggestion that as we were using bottle-top ingots they could be made higher than ordinary rimming steel, and we are now trying 105-in. bottle top ingots, and have reduced both width and thickness.

The estimated rolling time for 60,000 tons on the present high-lift slabbing mill is 132 h. This is the time which would be taken to roll with no mistakes and no waiting for heat or maintenance. It is estimated that the universal slabbing mill will roll this same tonnage and same assortment of orders in 108 h. There is less likelihood of mistakes being made on the universal mill and there is no doubt that it will be easier to achieve the tonnage we required in 108 h on the universal mill than 132 h on the high-lift mill.

N E W S

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

Annual General Meeting, 1957

The Eighty-Eighth Annual General Meeting of The Iron and Steel Institute will be held at 4 Grosvenor Gardens, London, S.W.1, on Wednesday and Thursday, **22nd and 23rd May, 1957.** The programme of the meeting was published in the April issue, p. 553. The Dinner for Members will be held at Grosvenor House, Park Lane, London, W.1, on the evening of Wednesday, 22nd May.

NEWS OF MEMBERS

Mr. **F. A. Ball** has been awarded an Insignia in Technology by the City and Guilds of London Institute.

Mr. **R. W. Berry** has resigned his position as Technical Secretary of the Plant Engineering Division of B.I.S.R.A. to join the Development Section of the British Iron and Steel Federation.

Mr. **D. S. Box** has left the Plant Engineering Division of B.I.S.R.A.

Mr. **D. Bruce-Gardner** has been appointed to the Board of Guest Keen Iron and Steel Co., Ltd.

Mr. **D. Howell Davies** has left B.I.S.R.A. Sketty Hall Group to take up an appointment as Assistant Lecturer at the Swansea Technical College.

Mr. **W. G. Dilley** has been appointed a Director of George Elliot and Co., Ltd.

Mr. **M. F. Dowding** has been appointed a Director of Davy and United Roll Foundry Ltd.

Mr. **S. W. Evans** has been appointed to the Board of Guest Keen Iron and Steel Co., Ltd.

Mrs. **J. M. Frankau** is now Technical Secretary of the Plant Engineering Division of B.I.S.R.A.

Mr. **W. F. Gilbertson** has been elected President of the Iron and Steel Trades Employers' Association.

Mr. **D. Gilbey** has left the Ashanti Goldfields Corporation Ltd. to join Rhoanglo Mine Services, Kitwe, Northern Rhodesia.

Monsieur **P. Gouge** has retired from the Société Métallurgique de Normandie.

Mr. **P. R. Gray** has resigned his position as Manager of the Seraphim Ore Preparation Plant at Appleby-Frodingham to join the Dravo Corporation, Pittsburgh, U.S.A., as a sinter plant and ore preparation plant engineer.

Dr. **J. Heslop** has left the Naval Construction Research Establishment to join the Development and Research Department of the Mond Nickel Co., Ltd.

Mr. **M. M. Jones** is now employed as a graduate apprentice by Marconi Instruments Ltd., St. Albans.

Mr. **H. J. Miller** has left British Insulated Callender's Cables Ltd. to join I.C.I. Ltd. Metals Division.

Mr. **James H. Moore** is now Assistant to the President of Metal Hydrides Inc., Beverley, Mass., U.S.A.

Mr. **J. Moran** has been appointed a Director of The North Eastern Iron Refining Co., Ltd., and of its associated company, Stillington Estates Ltd. He will continue to be General Manager of both companies.

Mr. **G. R. Morton** has been appointed Senior Lecturer in Metallurgy at the Wolverhampton and Staffordshire College of Technology.

Mr. **E. Julian Pode** has been appointed a Director of Ore Carriers Ltd.

Dr. **F. D. Richardson** has been appointed Professor of Extraction Metallurgy at the Imperial College of Science and Technology.

Mr. **A. M. Sage** has been appointed Manager of the newly formed Development Department of the British Electro Metallurgical Company. He will continue to be responsible for a period for B.I.S.R.A.'s work on structural steels in a consultative capacity.

Mr. **B. A. Sargent** has been appointed Chief Resident Representative of The Wellman Smith Owen Corporation Ltd. in Spain.

Mr. **A. W. Stones** is now District Manager of the Cardiff branch of British Oxygen Gases Ltd.

Mr. **A. P. T. Taylor Gill** of Edgar Allen and Co., Ltd., has returned to the U.K. to be Technical Representative for tool steels in the North of England. He completes 30 years with the Company in September.

Mr. **J. S. Walton** is now at the Wern Works of High Duty Alloys Ltd., Briton Ferry.

Obituary

Mr. **William Tait Cowan** (*elected* 1951), of Musselburgh, on 14th August, 1956.

Mr. **Rowland Staton** (*elected* 1953), of Sheffield, on 13th February, 1957.

Monsieur **F. A. Vandestrick** (*elected* 1952), of Seraing, Belgium, on 25th August, 1956.

Mr. **Bernard Wragg** (*elected* 1910), of Loxley, near Sheffield, on 2nd February, 1957.

EDUCATION

Cambridge Summer School

A Summer School on "The Use of Electrons in the Examination of Metals" is being held in the Departments of Metallurgy and Physics at Cambridge University from 9th to 20th July, 1957. The subjects covered will include recent advances in electron microscopy as applied to the examination of metals, scanning electron microscopy, selected area diffraction, and X-ray microscopy and micro-analysis. The fee for the course is £25; accommodation will be provided at Pembroke College. Further information and application forms may be obtained from the Secretary, Board of Extra-Mural Studies, Stuart House, Mill Lane, Cambridge.

Technical State Scholarships 1957

A number of Technical State Scholarships, tenable at universities, university colleges, and establishments for further education, will be awarded in 1957 to students who have been in attendance at establishments for further education. Applications should be submitted through the Principal of the establishment the candidate is attending.

Joint Committee for National Certificates in Metallurgy

The Committee reports the following results of examinations held in 1956, and comparative figures:

Ordinary National Certificate

Year	Technical Colleges	Entered	Total Certificates awarded	With Distinction	With Prize Award
1946	4	28	23	4	7
1948	15	123	74	12	16
1950	20	281	142	23	26
1952	28	334	186	22	31
1954	37	522	258	25	39
1955	37	523	275	33	38
1956	**37**	**572**	**196**	**31**	**42**

Entered since 1946: 3410. Certificates awarded: 1752

Higher National Certificate

1946	—	—	—	—	—
1948	8	38	22	1	2
1950	11	62	42	8	10
1952	16	132	120	16	22
1954	18	173	144	22	31
1955	21	183	138	22	30
1956	**23**	**247**	**205**	**23**	**27**

Entered since 1946: 1181. Certificates awarded: 918

Endorsement Subjects—Ordinary National Certificate

1955	5	59	27	1	1
1956	**6**	**56**	**23**	**7**	**7**

Endorsement Subjects—Higher National Certificate

1955	11	70	66	7	7
1956	**12**	**102**	**89**	**8**	**8**

Assessed Examination in Mathematics S.2

Year	Technical Colleges	Candidates Entered	Passed Examinations
1952	5	77	44
1953	11	169	122
1954	22	321	220
1955	23	487	284
1956	**28**	**570**	**391**
		1624	1061

Of the successful candidates for the Higher National Certificate in Metallurgy in the five years 1952–1956, 92 attained the exemption standard of The Institution of Metallurgists and qualified for exemption from the Licentiateship examinations in three or more subjects.

AFFILIATED LOCAL SOCIETIES

New Cleveland Award

Under the terms of the will of the late Mr. T. H. W. Chambers, a sum of £50 was left to the Cleveland Institution of Engineers to encourage younger members to submit papers. The Council have decided to award annually a prize of £20 in books or cash for the best paper on a subject acceptable to the Institution, which is submitted by a member not more than 30 years of age on 31st December of the year in which the paper is submitted. Papers should be submitted to the Secretary (F. G. E. Webster, Esq.), Cleveland Institution of Engineers, 12 South Park Avenue, Normanby, Middlesbrough, by not later than 30th September. Three copies of the paper, which should be between 6000 and 10,000 words long, should be submitted, together with an application form which is obtainable from the Secretary of the Institution.

NEWS OF SCIENCE AND INDUSTRY

Conference on Quality Measurement

The Society of Instrument Technology is holding a Conference on Automatic Measurement of Quality in Process Plants at University College, Swansea, from 23rd to 26th September. The main object of the conference is to enable users and makers of automatic measuring equipment and those concerned with research and development in this field to meet and discuss their work. The range of topics to be considered will include chromatography, mass spectrography, pH measurement, spectroscopy, titration, moisture content measurement, and viscosity and consistency determination. Further details may be obtained from the Secretary of the Society, 20 Queen Anne Street, London W.1.

United Steels' Walking Dragline

A W 600 walking dragline is now in use for removing the overburden at the Crosby Warren ironstone quarry of the Ore Mining Branch of The United Steel Companies Ltd. at Scunthorpe. The dragline has a maximum capacity of 10 yd^3, and a 184 ft dumping radius, with a boom length of 186 ft. It can remove about 400 yd^3 of spoil per hour.

Chemical Engineering Congress

The 12th Chemical Engineering Exhibition and Congress and the 2nd Congress of the European Federation of Chemical Engineering are being held concurrently at Frankfurt-am-Main from 31st May to 8th June, 1957. The organization has been in the hands of the Deutsche Gesellschaft für Chemisches Apparatewesen (DECHEMA).

Armstrong Whitworth Silver Jubilee

The 25th anniversary of the foundation of the Roll Division of Armstrong Whitworth (Metal Industries) Ltd. was observed on 26th February, 1957, when a dinner was held at Newcastle-upon-Tyne, attended by directors and executives of the Division and their wives.

Congress on Analytical Chemistry in Industry

The Scottish Section of the Society for Analytical Chemistry is organizing a Congress on Modern Analytical Chemistry in Industry at St. Andrews University from 24th to 28th June, 1957. The work of the congress will be divided into three sections: analysis in modern industry, the application of some newer analytical techniques in industry, and developments in analysis for new problems in industry. The speakers will include Mr. B. Bagshawe, of Thos. Firth and John Brown Ltd., and Mr. D. Manterfield, of Steel Peech and Tozer. The Congress is already fully subscribed, but it is hoped to publish the proceedings in due course.

NIFES Refresher Course

The Summer School of the National Industrial Fuel Efficiency Service this year will be held at Prestatyn, North Wales, from 3rd to 8th June. The theme will be "Routine Testing and Operation in Relation to Factory Fuel Consumption." Applications for admission to the course should be sent to NIFES, Baltic House, Mountstuart Square, Cardiff.

New Blast-furnace for Dorman Long

The second of the two new blast-furnaces at the Clay Lane Ironworks of the Cleveland Works of Dorman Long (Steel) Ltd. was lit recently. Each of the new furnaces is of 27 ft 6 in. hearth dia., and they are each expected to produce 1000 tons of pig iron daily. The

first furnace was brought into operation in December, 1956.

Mains-frequency Coreless Induction Furnace

Britain's first main-frequency coreless induction furnace, of 5 tons capacity, was recently installed at International Combustion Ltd., Derby. The furnace, which was built and installed by Birlec Ltd., is being used principally for spheroidal graphite iron, but this type of furnace has a wide field of applications.

Graphite Resistor Furnace

The Electric Furnace Co., Ltd., have recently installed the first graphite resistor furnace for steelmaking in the U.K. at the works of Lake and Elliot Ltd., Braintree. This 15-cwt 350-kVA furnace is being used for carbon and low-alloy steels. Modifications in the design of the standard type of furnace have been made, to permit scrap charging, fettling of the lining, and slag removal, which were difficult with earlier designs.

New A.S.T.M. Publications

Among recent publications of the American Society for Testing Materials are a Report on Elevated-Temperature Properties of Wrought Medium-Carbon Alloy Steels (Special Technical Publication 199: $4.25) and a Symposium on Properties, Tests, and Performance of Electrodeposited Metallic Coatings (Special Technical Publication 197: $3.00). The Society also offers three series of punched-card indexes to absorption spectral data: an Empirical Formula Name Index for Infra-red Spectra, a Spectral Absorption Index for Infra-red Spectra, and a Spectral Index for Ultra-violet Absorption Spectra. Full information can be obtained from the Society, 1916 Race Street, Philadelphia 3, Pa., U.S.A.

Open Day at Fulmer Institute

On the tenth anniversary of the opening of the Fulmer Research Institute by the late Sir Stafford Cripps, an Open Day is being held on 2nd July. The laboratories will be available for inspection and some aspects of the work of the Institute will be exhibited.

New Plant at Appleby-Frodingham

During the last week of 1956, a new turbo-blower and open-hearth furnace were brought into operation at the Appleby-Frodingham Steel Company, Scunthorpe. The turbo-blower is capable of delivering 150,000 ft^3 of air per minute at 35 lb/in^2, and will help in increasing the pig-iron production of the blast-furnaces. The new 300/350-ton tilting furnace is the third to be installed in the Frodingham melting shop. Its output will largely be used in the new Brinsworth mill of Steel, Peech, and Tozer.

Tin Researches in 1956

The Annual Report of the International Tin Research Council for 1956 has recently been published. Of especial interest is an account of the new aluminium–tin bearing alloy, which can now be bonded to steel strip more easily than hitherto. Some results have also been obtained which suggest that 'bright' tin coatings by electrodeposition will soon be possible.

Head Wrightson Changes

The Iron Foundries Division of Head Wrightson has become a wholly owned subsidiary of Head Wrightson and Co., Ltd., with the title of Head Wrightson Iron Foundries Ltd. A new wholly owned subsidiary has been formed, to be known as Head Wrightson Iron and Steelworks Engineering Ltd.; it will handle the work of the Company's former Iron and Steel Division, and will also cover a wider range of engineering and contracting for iron and steelworks construction.

The P.T. Oxygen Gun

The Steel Company of Wales Ltd. have recently concluded an agreement with Kaiser Engineers of Oakland, Calif., U.S.A., in respect of the P.T. oxygen gun, giving them full rights of manufacture, sale, etc. throughout the world, with the exception of Europe. The P.T. oxygen gun was developed by the Company for use in open-hearth furnaces for rapid carbon elimination by jet impingement of oxygen on to the surface of the steel by means of a gun centrally mounted through the furnace roof. It is in regular use at the Abbey Works melting shop at Margam, delivering over 60,000 ft^3/h of oxygen per furnace without interference to charging or tapping.

British Standard for Springs

British Standard 2803 : 1956 *Oil hardened and tempered steel wire for springs* has recently been published at 3*s*. It deals with three grades of wire: Grade 1, high-duty (ground); Grade 2, high-duty; and Grade 3, commercial. Requirements are given for tensile, torsion, reverse-bend, wrapping, deep-etch, decarburization, and dead-wire tests.

Vacu-Blast in France

Vacu-Blast Limited, manufacturers of mobile shot-blast machinery, have established a company in France, known as Société Vacu-Blast (France), with offices at 46 rue Anatole France, Levallois-Perret (Seine), France.

Iron Ore Terminal

A project for the building of a new iron-ore terminal near Milford Haven is under consideration by the Steel Company of Wales Ltd. and Guest Keen Iron and Steel Co., Ltd. New deep-water unloading facilities are needed to receive the large ore carriers now being built for the iron and steel industry.

New Division for Birlec

A new Heating Division has been formed by Birlec Ltd., of Erdington, Birmingham, resulting from the amalgamation of the former Furnace and Induction Heating Divisions. Manager of the new Division is Mr. J. M. Paton.

Colour Photography in Industry

An exhibition of colour photographs by Adolf Morath was held recently at the Tea Centre, Lower Regent Street, London, S.W.1. On display were some magnificent photographs of steelmaking processes and works, several of which have been featured on the cover of *Steel Review*, published by the British Iron and Steel Federation.

Gamma-Radiography Exposure Time Calculator

The British Steel Castings Research Association have modified the iridium scale of the slide rule which they supply for the calculation of exposure times in gamma radiography. The change has been made necessary by the decision of the Isotope Division of the Atomic Energy Research Establishment to revise the calibration figures for iridium sources which they supply. As from 1st January, 1957, these sources are being rated at 54% of their previous value in millicuries.

New slide rules embodying the revised iridium scale are now available from the Association at East Bank Road, Sheffield, 2, and existing users of the calculator can have a corrected slide fitted on payment of a small charge.

New Electric Furnace

The English Steel Corporation Ltd. have ordered from Birlec Ltd. a new electric-arc furnace for alloy-steel production at their Openshaw Works. The new furnace will have a shell diameter of 15 ft, giving a nominal charge capacity of 30 tons. A transformer rated at 10,000 kVA and fitted with on-load tap-changing gear will give the furnace a high melting rate.

International Mineral Dressing Congress

An International Mineral Dressing Congress is to be held at Stockholm from 18th to 21st September, 1957. Excursions to mining districts have been arranged to follow the meeting. Details can be obtained from the General Secretary, Näckströmgatan 1III, Stockholm C, Sweden.

Accident Prevention in the Iron and Steel Industry

The British Iron and Steel Federation have recently published three booklets on safety, entitled " The Safe Operation of Bar Reeling and Straightening Machines ", " The Prevention of Gassing Accidents in Iron and Steelworks ", and " The Safe Coupling of Ingot Casting Cars and Scrap Pan Bogies ". Copies are available from Steel House, Tothill Street, London, S.W.1.

Electron Microscope for Swinden Laboratories

A Siemens II electron microscope has been installed at the Research and Development Department of the United Steel Companies Ltd. in Rotherham.

Institution of Chemical Engineers

A Royal Charter has been granted to the Institution of Chemical Engineers.

Symposium on Sinter

A Symposium on Sinter is being held in Paris by the Institut de Recherches de la Sidérurgie from 21st to 23rd May, 1957. Full details were given on p. 529 of the April issue of the *Journal*.

Industrial Publications Received

" *Chemical Service in Defence of the Realm* ", published by the Ministry of Supply, is a history of the Chemical Inspectorate during the past hundred years.

Marco Conveyor and Engineering Co., Ltd., have published a brochure entitled " *Elevators and Skip Hoists* ".

The Fusarc/CO_2 Automatic Arc Welding process, recently introduced by Quasi-Arc Ltd. of Bilston, is described in technical circular T.C.107. It is claimed that the new process has major applications in the shipbuilding and automobile industries.

Publication No. 68 of W. C. Holmes and Co., Ltd., Huddersfield, describes the Holmes–Rothemuhle Multi-Cell Cyclone Dust Catcher.

Newton, Chambers and Co., Ltd., have produced an illustrated 20-pp. brochure dealing with iron and steelworks equipment. It covers a wide range of blast-furnace and steelworks ancillary plant.

Vitreosil electric immersion heaters are described in a brochure from the Thermal Syndicate Ltd.

Some of the products of the High Vacuum Engineering Division of W. C. Heraeus G.m.b.H. of Hanau, W. Germany, are listed in a new catalogue obtainable from their U.K. agents, Fleischmann (London) Ltd., 16 Northumberland Avenue, London, W.C.2.

The Hilger Fluorite Polychromator, which can be used for the rapid determination of carbon, phosphorus, and sulphur in steel, is the subject of catalogue E. 796, obtainable from Hilger and Watts Ltd., 98 St. Pancras Way, London, N.W.1.

Corrigendum

In Part I of the paper by K. E. Puttick entitled " The Structure, Deformation, and Fracture of Pearlite " (*Journal*, 1957, vol. 185, February pp. 161–176), the captions to Figs. 8 and 10 were interchanged. The captions should have read:

Fig. 8—Onset of rod-like growth in pearlite colony
Fig. 10—Branching of cementite (replica of deformed surface)

DIARY

1st May—Institute of Metals (in conjunction with British Nuclear Energy Conference)—Half-day discussion on " The Metallurgy of Niobium "—Church House, Westminster, London, S.W.1, 10 a.m.

1st May—Institute of Metals and Institute of British Foundrymen—Symposium on Metallurgical Aspects of the Control of Quality in Non-Ferrous Castings—Church House, London, S.W.1, 9.30 a.m.

2nd May—Institution of Civil Engineers—Annual Dinner—Dorchester Hotel, Park Lane, London, W.1.

2nd May—Leeds Metallurgical Society—" Some Aspects of Creep," by J. D. Meakin—Large Chemistry Lecture Theatre, The University, Leeds, 7.30 p.m.

6th–17th May—British Industries Fair—Birmingham.

7th May—Society of Chemical Industry (Corrosion Group)—Annual General Meeting and Presidential Address " A Pilgrim's Progress from Corrosion," by T. Henry Turner—14 Belgrave Square, London, W.1, 6.30 p.m.

7th–17th May—Instruments, Electronics, and Automation Exhibition—Olympia, London.

8th and 9th May—British Iron and Steel Research Association—48th Steelmaking Conference—Ashorne Hill.

19th–23rd May—European Investment Casters Association—Annual Conference—Stratford-on-Avon.

21st–23rd May—Institut de Recherches de la Sidérurgie—Symposium on Sinter—Maison de la Chimie, 28 rue St.-Dominique, Paris 7.

22nd–23rd May—American Iron and Steel Institute—Annual Meeting—New York.

22nd–23rd May—The Iron and Steel Institute—Annual General Meeting—4 Grosvenor Gardens, London, S.W.1.

22nd May—The Iron and Steel Institute—Dinner for Members—Grosvenor House, Park Lane, London, W.1, 7.0 for 7.30 p.m.

28th and 29th May—Het Koninklijk Instituut Van Ingenieurs, de Koninklijke Nederlandse Chemische Vereniging, The Society of Chemical Industry, and The Institution of Chemical Engineers—Joint Symposium on the Scaling-up of Chemical Plant and Processes—Church House, London, S.W.1.

31st May–2nd June—National Industrial Safety Conference—Scarborough

31st May–8th June—12th Chemical Engineering Exhibition and Congress and 2nd Congress of the European Federation of Chemical Engineering—Frankfurt-am-Main, W. Germany.

TRANSLATION SERVICE

(The previous announcement was made in the April, 1957, issue of the *Journal*, p. 556).

Translations Available

No. 523 (French). J. Calmettes and J. Drain: " The Determination of Nitrogen in Steel by a Wet Method." (*Revue de Métallurgie, Mémoires*, 1956, vol. 53, Sept., pp. 682–688; Discussion, p. 688).

No. 524 (German). F. Harders, H. Knüppel, and K. Brotzmann: " The Industrial Large-scale Treatment of Steel Melts under Reduced Pressure." (*Stahl und Eisen*, 1956, vol. 76, Dec. 27, pp. 1721–1728).

No. 525 (German). E. Wilms: " What the Rolling Mill Engineer Requires of the Electrical and Mechanical Part of Single-stand Reversing Mills." (*Stahl und Eisen*, 1956, vol. 76, Feb. 23, pp. 185–192).

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

CONTENTS

	PAGE
Mineral Resources	117
Ores—Mining and Treatment	117
Fuel—Preparation, Properties and Uses	118
Air Pollution and Smoke	118
Temperature Measurement and Control	118
Refractory Materials	119
Iron and Steel, General	119
Blast-Furnace Practice and Production of Pig Iron	119
Treatment and Use of Slags	120
Production of Wrought Iron	120
Production of Steel	120
Production of Ferro-Alloys	122
Foundry Practice	122
Vacuum Metallurgy	126
Reheating Furnaces and Soaking Pits	126
Heat-Treatment and Heat-Treatment Furnaces	126
Forging, Stamping, Drawing and Pressing	128
Rolling Mill Practice	128
Machinery for Iron and Steel Plant	129
Lubrication	129
Welding and Flame-Cutting	130
Machining and Machinability	130
Cleaning and Pickling	131
Protective Coatings	132
Powder Metallurgy	134
Ferrites, Cermets and Carbides	135
Properties and Tests	135
Metallography	143
Corrosion	145
Analysis	146
Industrial Uses and Applications	148
Historical	149
Economics and Statistics	150
Miscellaneous	150
Book Notices	150
New Publications	152

MINERAL RESOURCES

The Geology of the East Midlands Ironstone Fields. 1956, **22**, Aug., 322–328; Sept., 372–375). The Abbott Memorial Lecture at Nottingham.

Geology and Mineral Resources of the Ivanpah Quadrangle, California and Nevada. D. F. Hewett. (*U.S. Geol. Survey, Prof. Paper*, 275, 1956, pp. 172). Iron deposits are recorded.

Geology and Ore Deposits of the Zimapán Mining District, State of Hidalgo, Mexico. F. S. Simons and E. Mapes. (*U.S. Geol. Survey, Prof. Paper*, 284, 1956, pp, 128). Haematite and limonite occurrences are noted.

Geology and Economic Aspects of the Manganese Ore Deposits of Jamda-Koira Valley, Keonjhar and Bonai, Orissa. B. B. Engineer. (*Tisco*, 1956, **3**, Oct., 202–210). The author describes the distribution, general geology, size and shape of the ore bodies and discusses their genesis and their structural relations to country rocks. The grades of ore are evaluated and the reserves estimated. A geological map and photographs of specimens are included.—L. E. W.

Grade and Depth of Manganese Ores of Kodur, Srikakulam District. J. S. R. Krishna Rao. (*J. Sci. Indust. Res.*, 1956, **15B**, July, 413). Analysis of samples showed that the percentage of manganese increased with depth, while the iron and silicon contents decreased and the phosphorus content was uniform.—E. E. W.

The Lateritic Iron Ore of Goa. J. Jungwirth. (*Berg. Hüttenmänn. Monatsh.*, 1956, **101**, Apr., 93–99). The geology of the area, nature of the deposits, mining methods and transport are reviewed.

ORES—MINING AND TREATMENT

The Lorraine Iron Mines. G. Pajot and H. Maria. (*Mine Quarry Eng.*, 1956, **22**, Apr., 126–135). Mining methods and modernization are described.

Recent Developments in Iron Ore Mining Outside Germany. E. Plotzki. (*Stahl u. Eisen*, 1956, **76**, Oct. 4, 1297–1301). The author discusses the increase in demand for iron ore in the world and gives output figures of new spheres of ore mining with regard to the future German supply of iron ore.—T. G.

The Kiruna Operations. C.-B. Berglund. (*Mine Quarry Eng.*, 1956, **22**, Jan., 2–10; Feb., 46–57). The history and geology of the (Swedish) iron ore area and the change over from open to underground mining are described.

Magnetite. (*Mine Quarry Eng.*, 1956, **22**, July, 270–271). An account of the mineral and its properties.

Raw Materials for the Steelworks at Jamshedpur. M. V. Wazalwar. (*Tisco*, 1956, **3**, Oct., 194–201). The sources of raw materials for the Tata Iron and Steel Works are described and shown on a part map of India. The general characteristics and average compositions of coal, iron ore, manganese ore, chromite, limestone, fireclay, etc. are given with data on annual consumption.—L. E. W.

Sampling of Minerals. Weights Suitable for Samples. Nomograms for Sampling. P. Gy. (*Rev. Ind. Min.*, 1956, **38**, Feb., 53–99). A comprehensive statistical account of ore sampling with many nomograms for optimum weight.

Contribution to the Study of Magnetic Ore Separation. R. Tillé and W. Kirkpatrick. (*Rev. Ind. Min.*, 1956, **38**, Feb., 7–20). Formulæ are presented and their applications to industrial separations explained.

Froth Flotations: A Half-century Review. P. F. Whelan. (*Indust. Chem.*, 1956, **32**, Aug., 315–318; Oct., 409–411; Nov., 489–491). A classified list of minerals separated in this way is given, with an historical review. (113 references).

Recovery of Pyrites from the Coal Washery Rejects from Nowrozabad, Vindhya Pradesh. S. K. Banerjee and P. I. A. Narayanan. (*J. Sci. Indust. Res.*, 1956, **15B**, July, 402–407). Owing to the shortage of sulphur in India, new sources have been investigated. Samples of rejects from the washery were crushed and jigged. Some were then subjected to tabling or to flotation. Results are given.—E. E. W.

Beneficiation Studies of Nickeliferous Ores from the Shamrock Mine, Jackson County, Oreg., and the Congress Mine, Ferry County, Wash. J. E. Shelton. (*U.S. Bur. Mines, Rep. Invest.*, 5261, 1956, Oct., pp. 8).

A Mineral-Dressing Study of Manganese Deposits of West-Central Arkansas. M. M. Fine and D. W. Frommer. (*U.S. Bur. Mines, Rep. Invest.*, 5262, 1956, Oct., pp. 21).

Micrographic Study of Magnetite Oxidation. I. Behar. (*Compt. Rend.*, 1956, **243**, Dec. 5, 1877–1880). After a period of induction α-Fe_2O_3 nuclei form and develop on the four planes (111) one of which appears to be of special importance in the oxidation process.

Ore Preparation Plant at Margam: Effect on Blast Furnace Performance. K. C. Sharp and H. R. Tufnail. (*Iron Coal Trades Rev.*, 1956, **173**, Oct. 19, 945–955). The authors describe the ore preparation plant at the Margam Works of

Steel Co. of Wales Ltd., giving details of the ore crushing and screening plant, ore fines stockyard, coke crushing and screening plant, the handling of the flue dust, and the sinter plant and its operation. The effect of ore preparation on blast furnace performance is evaluated, with attention given to iron quality, coke rate, and blast temperature and pressure.

Reaction Zones in Iron-Ore Sintering Process. R. D. Burlingame, G. Bitsianes and T. L. Joseph. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 17). An experimental sintering machine is described and partly sintered charges were examined zone by zone. A resin-impregnated charge system is also used. Five zones are distinguished and discussed.

Sintering Iron Ores and Concentrates at Extaca. R. L. Bennett, R. E. Hagen and M. V. Mielke. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw. Mat. Comm. Proc.*, 1956, preprint, pp. 19). The plant and process is briefly outlined and equipment specifications given.

Weirton's New 8-foot Wide Iron-ore Sintering Plant. J. H. Strassburger. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*. 1956, preprint, pp. 9).

Design and Operation of No. 3. Sinter Strand at Jones and Laughlin's Benson Mines. R. G. Fleck and F. M. Hamilton. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 5).

On Problems Related to the Production of Self-Fluxing Sinter Agglomerates. J. Duda. (*Hutnické Listy*, 1956, **9,** (8), 461–466). [In Czech]. A detailed study was made of agglomeration using Krivoi Rog iron ores, flue dust, blast-furnace dust and calcined pyrites. Sintering conditions leading to maximum sinter strength and highest sintering rates were determined. The favourable influence of limestone on the sintering rates was investigated, suitably granulated dolomitic limestone was also used. The function of these additions was established by mineralogical analyses of the sinters.—P. F.

On the Magnetising Roasting of the Ni–Cr Iron Ore in South Asia District by Fluidised Bed. T. Mitsuhashi, M. Ueno and M. Tanaka. (*Tetsu to Hagane*, 1956, **42,** June, 461–466). [In Japanese]. Details are given of an experimental fluidised-bed roasting method, used with a laterite containing nickel and chromium. The iron recoveries were 85–90% in the overflow product, and 75% in the cyclone dusts; the nickel recoveries were of the same order, but the magnetic separation of chromium was not good. The concentrates contained 50–60% of iron.—K. E. J.

Determining the Reducibility of Iron Ores. F. Wald. (*Hutnické Listy*, 1956, **9,** (9), 532–536). [In Czech]. Existing methods are reviewed, and details are given of an apparatus constructed by the author and working on the Kräber-Luyken principle.—P. F.

FUEL—PREPARATION, PROPERTIES, AND USES

Coal Sampling. A. Ghosal. (*J. Sci. Indust. Res.*, 1956, **15A,** Aug., 363–368). Reasons are given for choosing a 9 lb scoop as " optimum " for drawing samples from wagons during loading or unloading.—E. E. W.

The Buoyancy Effects in Furnaces and Boilers. G. Horn and M. W. Thring. (*J. Inst. Fuel*, 1956, **29,** Oct., 437–440). Model and theoretical work is described to show the effect of buoyancy in furnace systems; applications of the results to practical cases is shown.—D. L. C. P.

The Supply of Coal for Coke Manufacture. R. Lowe. (*Colliery Guardian*, 1956, **193,** Dec. 6, 679–684). Requirements, coal types and reserves are reviewed as well as blending and sulphur removal. The Gray-King test is described and coal ranking based upon it.

Injury Experience in the Coking Industry, 1954. S. T. Reese and N. W. Kearney. (*U.S. Bur. Mines, Bull.*, 564, 1956, pp. 18). Tables analysing the data are given.

A Study of Pressure Relationships in Commercial Coke Ovens. D. C. Coleman and P. J. Farley. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 13). Wall pressures were measured in a test oven and gas pressures in a commercial oven with 21 coal blends all of which developed two pressure peaks. Conclusions are indicated.

Studies in the Mechanism of Coke Formation: Part II—Plastic Properties, Rank and Coking Characteristics of Coals. S. R. Ghosh, N. N. das Gupta and A. Lahiri. (*J. Sci. Indust. Res.*, 1956, **15B,** Sept., 518–529). The flow properties of different Indian coals in the carbonization temperature range of 300°–500° C, determined by a Gieseler plastometer, were correlated with rank of coal. The physical properties of cokes, as indicated by Shatter, Micum and Haven tests, were found to be related to the plastic properties.—E. E. W.

Coke Oven to Measure Expansion Pressure—Modified Illinois Oven. H. W. Jackman, R. J. Helfinstine, R. L. Eissler, and F. H. Reed. (*Amer. Inst. Min. Met. Eng., Blast Furn. Coke Oven Raw Mat. Comm.*, 1955, **14,** 204–219). The Illinois State Geological Survey has built an electrically-heated 17 in. movable-wall coke oven of 700 lb capacity to study the pressure exerted on oven walls during carbonisation. The design and operation of the oven are described in detail and the results obtained are discussed. In general, pressure measurements are about 20% lower than those obtained in other movable-wall ovens, and coke properties and yields obtained correlate well with those obtained with similar blends in commercial ovens.—G. F.

The Manufacture of Cold Clean Producer Gas. W. Scott. (*Brit. Coal Utilisation Res. Assoc. Monthly Bull.*, 1956, **20,** Jan., 1–8). A review of washing systems and of the uses of coke and anthracite are given. (37 references).

Mixing Mechanisms in Gas Fired Combustion Chambers. F. Schoppe. (*VDI-Forschungsheft*, 456, 1956, pp. 39). The combustion of gases in chambers is treated theoretically and experimentally. Relations between flame characteristics, such as flame length and rate of heat release, and air velocity are obtained. The dependence of these relations on combustion chamber dimensions is stressed.—J.R.P.

AIR POLLUTION AND SMOKE

Heat Treating and the Smog Problem. S. S. Griswold. (*Metal Treatment*, 1956, **7,** July-Aug., 14–16, 40). The author cites a number of regulations and discusses the lessons learned by heat treaters in meeting them. The problem of smoke created by the introduction of oily parts into furnaces is emphasised.—A. H. M.

Chemical and Physical Particle Conductivity Factors in Electrical Precipitation. H. J. White. (*Chem. Eng. Prog.*, 1956, **52,** June, 244–248). Resistivity measurement, conditioning by humidification, chemical and other means and surface conduction studies are discussed, and values given.

TEMPERATURE MEASUREMENT AND CONTROL

Pyrometers. L. Neall. (*Inst. Vitreous Enamellers Bull.*, 1956, **7,** Sept., 7–8, 10). The construction, installation and maintenance of thermocouple pyrometers is described.—A.D.H.

British Pyrometer Progress. (*Brit. Steelmaker*, 1956, **22,** Dec., 358–360). The Land total radiation pyrometer is described and its calibration and the special furnace used discussed. Its mounting and use is then outlined.

A Thermocouple for High Temperatures. Advantages of the 'Five-Twenty' Couple. J. C. Chaston. (*Platinum Metal Review*, 1957, **1,** Jan., 20–22). The couple is made up of 5% and 20% Rh/Pt alloys. It may be used constantly up to 1700° and occasionally up to about 1825° C.

The Measurement by means of a Probe Radiometer of Radiant Heat Flux Distribution at Curved Surfaces. G. S. Cantle, M. N. El Din and M. F. Sakkal. (*J. Inst. Fuel*, 1956, **29,** Aug., 327–332). Measurements on heated tubes with different spacings are described.

Photoconductive Cells for Use in Radiation Pyrometry. A. M. Godridge. (*Brit. Coal Utilisation Res. Assoc. Monthly Bull.*, 1956, **20,** Sept., 349–355). Types are briefly reviewed and the theory of radiation pyrometry outlined.

The Dichromatic Radiation Pyrometer. W. S. Tandler, R. H. Jourin and M. Grossman. (*PB.* 116183, 1954; *TIDU List* 609). A prototype instrument is described and the theory given.

Experience in the Control of Liquid-Steel Temperatures with Thermocouples. S. G. Otlivanov and I. A. Sokolov. (*Stal'*, 1956, (5), 409–415). [In Russian]. The results of prolonged investigations at the Kuznetsk Metallurgical Combine on the use of immersion thermocouples for the determination of steel temperatures in the O.H. and electric furnaces and in ladles are presented and discussed. The design used secured a life of 18–20 measurements, the proportion of measurements with error over ± 10–15% being 3–5%. Based on comparisons of steel quality with results of temperature determinations, recommended temperature ranges for different stages of

melting for various steels are listed. Temperature control by regulation of charging and the flow of coke-oven gas, was effective.—S. K.

REFRACTORY MATERIALS

Modern Trends in Fireclay Refractories in Iron and Steel Works. H. R. Lahr. (*Refract. J.*, 1956, **32,** July, 316–330). **Discussion** (332–336). Recent experience in the use of refractories throughout iron and steel works is described. It has been found that the use of high-alumina air setting cement gives outstanding results when used for jointing in 'Kling' blast furnace ladles. Experiments comparing the stiff plastic process with the semi-dry process for making low porosity fireclay bricks are explained. The following are among subjects raised in discussion: blast furnace hearth construction and lining materials; open hearth checkers; coring.—D. L. C. P.

Refractory Practice and Mineral Dressing. Developments in Hungarian Methods and Equipment. M. Rácz. (*Refract. J.*, 1956, **32,** Oct., 494–504). The most recent practice in refractory production is described, and an example of the flow arrangements for modern fireclay and silica brick production is given.—D. L. C. P.

Study of Indian Chrome Ores for the Refractories Industry: Part I—Mineralogy and Origin of Indian Chromites. M. Ramakrishna Rao, V. K. Moorthy and Rabindar Singh. (*J. Sci. Indust. Res.*, 1956, **15B,** Aug., 455–466). The mineralogy of 16 chrome ores has been studied by microscopical methods and differential thermal analysis. All appear to be of magmatic origin.—E. E. W.

Production of Forsterite Refractories: Part III—Effect of Grain Size and Precalcination on Magnesite-talc Schist-magnesite Compositions. Sudhir Sen and Rabindar Singh. (*J. Sci. Indust. Res.*, 1956, **15B,** July, 383–386). The addition of calcined magnesite to schist effectively decreased the porosity and increased the crushing strength of the refractories. Reduction in the particle size of the schist reduced the porosity irrespective of composition. Pinholes were eliminated by pre-calcination at 1,550° C or by fine crushing (— 100 mesh).

Treated Silica Brick Eliminates Dust. C. E. Cline and R. A. Bloom. (*Iron Age*, 1956, **178,** Aug. 23, 100–101). The use of ammonium lignosulphonates to dust-proof silica bricks is described. The solution is applied by dipping or spraying and eliminates dust during handling. There is no effect on refractory quality.—D. L. C. P.

Observations on the Stabilization of Zirconia. B. C. Weber, H. J. Garrett, F. A. Mauer and M. A. Schwartz. (*J. Amer. Ceram. Soc.*, 1956, **39,** June, 197–206). The characteristics of magnesia-, lime- and ceria-stabilized zirconia were determined by X-ray analysis. Also investigated were the systems TiO–ZrO_2 and Ti–ZrO_2.—D. L. C. P.

The Measurement of the Thermal Conductivity of Refractory Materials. F. H. Norton and W. D. Kingery. (*Ceramics*, 1956, **8,** April, 10–18). The physical theory of thermal conductivity of single phase ceramic materials is considered and compared with experimental results. The variation with temperature is particularly considered.—D. L. C. P.

An Experimental Study of the Thermal Conductivity of Some Egyptian Fireclay Insulating Bricks. I. I. Sheriff and M. Y. Bakr. (*Silicates Indust.*, 1956, **21,** June-July, 287–289). [In English].—P. F.

Thermal Expansion of Various Ceramic Materials to 1500° C. O. J. Whittemore jun. and N. N. Ault. (*J. Amer. Ceram. Soc.*, 1956, **39,** Dec., 443–444). Apparatus is described and results for 21 materials tabulated.

The Mechanical Strength of Silica and Basic Bricks. G. R. Rigby and W. R. Davis. (*Refract. J.*, 1956, **32,** Oct., 482–492). A sonic method of measuring Young's Modulus of refractory materials is described: its advantages include—accuracy $\pm$ 4%; value is average for specimen; specimen unaltered by determination. Experiments are described which lead to the conclusion that the strength of silica bricks does not suffer by repeated heating to 1000° C; with chrome magnesite and magnesite-chrome bricks a progressive reduction in strength and load-bearing capacity can be anticipated during service.

Dynamic and Static Tests for Mechanical Properties of Fired Plastic Refractories and Other More Resilient Materials. L. E. Mong and W. L. Pendergast. (*J. Amer. Ceram. Soc.*, 1956, **39,** Sept., 301–308). An improved loading apparatus and a sensitive deflectometer for transverse tests are described. Dynamic moduli of elasticity were determined for 4 resilient materials and for 9 brands of heat treated plastic refractories.

Comparative Evaluation of Ladle Bricks. F. Gottardi and A. Palazzi. (*Met. Ital.*, 1956, **48,** March, 111–118). [In Italian]. Results are given of tests carried out on ladle bricks produced by 4 different manufacturers. The tests where constructed to assess the practical behaviour of the bricks under works conditions. Great care was exercised to establish reliable and strictly comparable data.—M. D. J. B.

The British Ceramic Research Association. A. E. Dodd. (*Metallurgia*, 1956, **54,** Oct., 161–162). The activities of this association are discussed. Work on ladle bricks which has shown the advantage of bricks which 'bloat' on heating is considered. The chemical composition of the brick and its refractoriness do not seem to be of importance.—B. G. B.

Comparative Study of Refractory Runner Bricks. Their Behaviour in Use. F. Gottardi and A. Palazzi. (*Met. Ital.*, 1956, **48,** June, 275–280). [In Italian] The authors describe full-scale trials carried out on pit runner and funnel bricks from three different manufacturers. The object of the trials was to assess the operational properties of the various bricks without interfering with normal production. Brick quality was evaluated by measuring wear after teeming. The results of the tests have been checked by statistical methods.

IRON AND STEEL—GENERAL

Ferrous Metallurgy in Siberia. (*Stal'*, 1956 (5), 387–388). [In Russian]. Measures by which the planned construction in Siberia of works for the production of 15–20 million tons of steel a year is to be realized are outlined, the fuel and ore resources to be used and the type of plant are discussed.

Russian Log. L. Wilson. (*Iron Steel Eng.*, 1956, **33,** Oct.) A brief illustrated account of a visit to the U.S.S.R.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

The Oldest Coke-Fired Blast-Furnaces Constructed in Kladno (Czechoslovakia) a Hundred Years Ago. F. Psota. (*Hutník*, 1956, **6,** (7), 222–224). [In Czech].—P. F.

Russian Blast-furnace Plant: Wide Use of Constant Humidity and High Top Pressure. (*Iron Coal Trades Rev.*, 1956, **173,** Sept., 28, 797–799). The use of constant-humidity blast and the extensive application of high top pressure have achieved improvements in blast-furnace operating in Russia. Details of recent work on these two aspects are given, with special reference to the hydrogen content of the product in the former case, and to gas cleaning in the latter case.—G. F.

A Comparison of Blast Furnace Penetration with Model Studies. J. B. Wagstaff and W. H. Holman. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 11). Models are described with more than one jet and the data are correlated with full scale studies. The relationship of penetration studies to roughness are not yet understood and are difficult to measure.

Methods for Increasing Production Rate of Flue-dust Sinter. E. C. Rudolphy and E. J. Whittenberger. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 7). Moisture control is described and the use of additives. Continuous measurement of permeability is required and this must be kept high by proper proportioning of the feed.

Axial Blast-Furnace Blower. (*Engineer*, 1956, **202,** Nov. 30, 783–784). An account of the Sulzer turbine-driven blower for Port Talbot.

Tests on a Low-Grade Acid Ore in the Neuves-Maisons Blast Furnace No. 7. M. Pasquet. (*Publ. Inst. Rech. Sid.*, 1956, Series A, (140), Sept., pp. 33). Compositions of Maron ores from the various levels are described in detail. The furnace and its auxiliary plant is shown in plans, and production and material and thermal balances given. Chapters are added on coke consumption, quality of pig iron and its variations in Si and S over a period of 25 days. It is shown that two of the ores cannot be use economically, being low in Fe and high in Si.

Operating Experiences with High Beneficiated Burdens. D. Joyce, W. P. Dowhaniuk and B. Marsden. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 6). Work at Algoma Steel, Ontario is reviewed over five years' working. Siderite and sinter are discussed and analyses given.

Rational Distribution of Materials in the Blast Furnace: Swedish Charging Methods for Improved Operation. (*Iron Coal Trades Rev.*, 1956, **173,** Oct. 12. 897–901). Orthodox methods of blast furnace charging are reviewed, and details are given of Swedish experimental charging equipment designed for automatic selective charging, distributing both large and small pieces with a high degree of uniformity throughout the burden. An account is given of the testing of the system on a normal furnace, the improved burden distribution leading to a 10% increase in production and a similar reduction in fuel consumption.—G. F.

Difficulties in Controlling Furnace Pressure. R. Jeschar. (*Stahl u. Eisen*, 1956, **76,** Oct. 4, 1284–1289). After discussing the importance of correct pressure in furnaces, experiments with a soaking pit are reported. The variation in pressure is attributed to too strong a draught from the exhaust fan and to a leakage in the butterfly valve in the exhaust system. Reduction in speed of the exhaust fan motor and of the leakage gap of the valve markedly improved the pressure fluctuations of the furnace.—T. G.

Experience in the Complete Automation of Charge Weighing in Blast-Furnace Plants. B. N. Zherebin and V. G. Gur'yanov. (*Stal'*, 1956, (5), 396–402). [In Russian]. Developments in the automation of coke and burden weighing at the Kuznetsk Metallurgical Combine since 1940 are described, with special emphasis on recent attempts at complete automation of weigh-car operation. A fairly satisfactory solution has been arrived at, but with the present multi-component burden the equipment is complicated. With conversion to sinter practice simplification would be possible. The advantages of placing all electrical equipment in the weigh-car cabin are discussed. With the design evolved, changes in the charging cycle are easily effected. Points requiring further study and the changes in personnel requirements due to automation of charging are described.—S. K.

Desulphurization of Blast Furnace Iron with Injected Calcium Carbide. D. E. Watkins and J. R. McFarland. (*Amer. Inst. Min. Met. Eng. Blast Furn. Coke Oven and Raw Mat. Comm. Proc.*, 1956, preprint, pp. 7). Method and effects of injection into the ladle are described. The process is efficient and reproducible.

Breakdown of Blast Furnace Scabs by Means of Explosives at ILVA Works, Piombino. D. Camici. (*Met. Ital.*, 1956, **48,** April, 160–170). [In Italian]. The author gives a detailed description of the removal by means of explosives of very severe scabs on the No. 2 blast furnace at the ILVA works. The methods of detecting the position of the scab, the nature of its chemical and mechanical compositions, the means by which explosives were introduced and detonated and the success of the operations generally are fully discussed.

Control of Gas Flow in the Blast Furnace. N. N. Chernov and I. F. Domnitskiĭ. (*Stal'*, 1956, (5), 402–408). [In Russian]. Examples for smoothly working furnaces of radial distribution of CO_2 at the stockline, wall temperatures, gas temperature in the various off-takes, stockline level, blast pressures, top-gas pressure and blast volume are given. The effect on these of the following conditions are described: peripheral working, centre working and channelling. The regular analysis of stockline gases along two mutually perpendicular diameters is recommended as an aid in controlling furnace working. Optimal working is obtained when the CO_2-content at the centre is 2–3% lower than at the walls and when the axis of the furnace is kept relatively open.—S. K.

Influence of Impurities and Role of Iron Carbides in Deposition of Carbon from Carbon Monoxide. T. F. Berry, R. N. Ames and R. B. Snow. (*J. Amer. Ceram. Soc.*, 1956, **39,** Sept., 308–318). The dissociation of carbon monoxide in contact with iron oxide was studied. From 400 to 565° C the catalyst is $Fe_{20}C_9$, from 565 to 700° C it is Fe_3C. Inhibiting and accelerating agents are considered. Some observations on used blast furnace fire-clay linings are made.—D. L. C. P.

Influence of Some Factors on the Size of the Oxidizing Zone. B. N. Zherebin, V. M. Minkin, I. D. Nikulinskiĭ, V. M. Obsharov, I. A. Suchkov and M. Ya. Ostroukhov. (*Stal'*, 1956, (5), 391–396). [In Russian]. The blast furnace oxidizing zone is defined as the space in front of the tuyeres with CO_2-content $> 2\%$. The influence of various factors on the size of this zone was studied over a number of years on furnaces with a useful volume of 1163 m³ and hearth diameter 7620 mm Increasing blast volume from 2275 to 2500 m³/min appreciably increased the size of the zone with or without control of blast moisture at 20 g/m³. Increases were also obtained with increasing blast moisture when the blast temperature was increased to compensate. Higher blast temperatures as such actually resulted in longer oxidizing zones and this is attributed to kinetic energy effects. Higher blast pressures led to shorter zones.—S. K.

TREATMENT AND USE OF SLAGS

Determination of 1600° and 1700° C Liquidus Lines in $CaO.2Al_2O_3$ and Al_2O_3 Stability Fields of the System CaO–Al_2O_3–SiO_2. F. C. Langenberg and J. Chipman. (*J. Amer. Ceram. Soc.*, 1956, **39,** Dec., 432–439). Saturated slags were analysed and the diagrams revised.

Activities in Ternary Silicate Melts. F. D. Richardson. (*Trans. Faraday Soc.*, 1956, **52,** Oct., 1312–1324). The FeO–CaO–SiO_2 system is considered and compared with MnO–CaO–SiO_2 and activity is related to cation size.

Phase Equilibrium Studies in the System Iron Oxide–Al_2O_3 in Air and at 1 Atm. O_2 Pressure. A. Muan and C. L. Gee. (*J. Amer. Ceram. Soc.*, 1956, **39,** June, 207–214). The quenching technique was used, from temperatures between 1085° and 1725° C. Stability ranges of the various phases were delineated, and approximate compositions of crystalline and liquid phases determined.—D. L. C. P.

WROUGHT IRON

The Use of German Silica-Rich Ores in the Bloomary Plant of the Maxhütte at Unterwellenborn. W. Kutzer. (*Stahl u. Eisen*, 1956, **76,** Oct. 4. 1301–1304). The lack of iron-rich ore and good coking coal in Eastern Germany led to the erection of a Krupp-bloomary plant at Unterwellenborn for use of ore containing up to 29% SiO_2 with iron 33% and Al_2O_3 10%. The process is described in detail and illustrated by tables, graphs, and heat-balance sheets. Up to 435 tons of ore per day were used according to its composition resulting in about 125 tons of puddled ball, requiring $8{\cdot}3 \times 10^6$ kg.cal per ton.—T. G.

PRODUCTION OF STEEL

Steel Production Today. R. L. Knight. (*J. Australian Inst. Met.*, 1956, **1,** Oct., 76–88). A review of recent furnace design, being the 5th Annual Lecture of the Institute.

Progress in the Iron and Steel Industry: Lines of Future Development. E. T. Sara. (*Iron Coal Trades Rev.*, 1956, **173,** Sept. 14, 671–677). The author discusses the rapid expansion of the British steel industry and shows how the present rate of progress compares with what is happening elsewhere in the world. The special problems of small plants and independent re-rollers are considered, and the lines of future development are outlined.—G. F.

Methods of Adding Oxygen for Oxidizing Impurities in the O.H. Bath. G. S. Sel'kin. (*Stal'*, 1955, (12), 1081–1089). [In Russian]. An account is given of large-scale tests of some old and new methods of introducing oxygen into the O.H. bath, with special reference to the location, design and life of the tuyeres and the violence of the interaction of jet and bath. Oxygen introduction via steel tubes through charging doors is not suitable for shops with many furnaces. The use of water-cooled jets situated above the slag during the blow and located in the back wall is limited by fuming and splashing and interference with crane operation and runner maintenance. The most promising way is to introduce the oxygen through the roof; special tuyeres can eliminate splashing and the process can be made completely automatic.—S. K.

Experience in the Conversion of Furnaces to Natural Gas. H. Siegers. (*Stahl u. Eisen*, 1956, **76,** June 28, 805–809). The properties of natural gas and the construction of piping systems are discussed. The conversion to natural gas made in certain German steelworks is described, and experience to date is reported.

Briton Ferry Steel Company: New 100-ton Steel Furnace. (*Brit. Steelmaker*), 1956, **22,** Sept., 256–258). Steelmaking capacity at Briton Ferry Steel Co. Ltd. is to be increased by the installation of a 100-ton fixed basic O.H. furnace of Maerz design, with all ancillary equipment. Brief details of the new installation are given.—G. F.

Comparison of Basic Roofs and Part Silica Part Basic Roofs. M. Mulsant. (*Centre Doc. Sidér. Circ. Inform. Tech.* 1956, No. 9. 1825–1829). The results of tests carried out on 70-ton O.H. furnaces at Mont-St.-Martin, France between 1949 and 1955 are described. The cost of basic roofs is 950 FF/dm²

and silica roofs 135FF/dm² so that there is a considerable financial incentive to use silica roofs. However, the productivity of all-basic roofed furnaces was higher than silica roofed furnaces and this is shown to be of considerably greater importance than any initial saving in the cost of the roof.

Large Charging Boxes and Fuel Modifications Key to Increased Steel Tonnage. (*Indust. Heating*, 1956, **23,** May, 1005–1006). The use of charging boxes having a capacity of 61 ft³ and atomisation of fuel oil with natural gas instead of steam has increased productivity at Granite City Steel Co., Ill.—A. D. H.

Equilibria of Sulfur and Oxygen Between Liquid Iron and Open Hearth-Type Slags. H. L. Bishop Jun., H. N. Lander, N. J. Grant, and J. Chipman. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8,** July, 862–868). The authors have combined the results of several independent studies of the equilibrium between molten iron and simple basic and acid slags. The effects of basicity and slag iron concentration on sulphur distribution and carbon and oxygen equilibrium contents is emphasized. The equilibrium between sulphur and oxygen is presented to show an overall reaction important in sulphur control, and a relationship between the limiting carbon content of the metal and the corresponding sulphur distribution ratio is determined.—G. F.

State of Oxidation of Open Hearth Slags. Physico-Chemical Study. G. Husson, B. Trentini and P. Vayssière. (*Centre Doc. Sidér. Circ. Inform. Tech.* 1956, **13,** No. 8, 1587–1598). The difficulties of the rapid determination of the oxygen content of the metal in the O.H. furnace and the relative ease of the determination of the oxygen content of the slag have led to a study of the relation between the two, and to factors which control the oxygen contents of slag and metal. B. G. B.

Application of Steam Fuel Ratio Control to Open Hearth Operations. W. H. Kahl. (*Iron Steel Eng.*, 1956, **33,** Aug., 72–77). Equipment and operation are described and advantages and disadvantages listed. The latter comprise cost of equipment and installation, time needed for training and greater demands on service and maintenance. The advantages are in the directions of better control and the improvements it brings.

Faster Open Hearth Charging: Improving Scrap and Supply Services. M. D. J. Brisby. (*Iron Steel*, 1956, **29,** May, 163–165). The author shows that speed of charging the O.H. furnace, with its definite effect on production rate, is largely governed by the scrap and supply services. He discusses methods of scrap preparation for improving scrap quality and increasing pan weight.—G. F.

Use of the Basic-Lined Hot-Blast Cupola in Open-Hearth Steelworks. (*Iron Coal Trades Rev.*, 1956, **173,** July 6, 47–50). A basic-lined hot-blast cupola has been installed as a pre-melting unit in the open-hearth plant of Walzwerke AG near Cologne. Details are given of the layout and operation of the cupola plant, and the effect of pre-melting on open-hearth operation is discussed. The economics of the system are outlined.—G. F.

Experience with Pen-Recording Exhaust Gas Analysers on the Siemens-Martin Furnace. H. Wolff. (*Stahl u. Eisen*, 1956, **76,** Jan. 26, 96–99). Reasons for oxygen analysis are given and the operation of the apparatus is described. Its use in the control of exhaust gas is indicated. The results obtained are discussed and economic advantages enumerated.

System for Continuous Measurement of Oxygen in Open Hearth Furnaces. (*Indust. Heating*, 1956, **23,** April, 780–786). A sample of flue gas drawn from the furnace by a steam injector is cleaned in a centrifugal separator and its oxygen measured and recorded in a magnetic meter. The system is used at the Jones and Laughlin Steel Corp., Pittsburgh.

Hot Blast Cupola in the Open Hearth Steel Works. S. Tunder. (*Mét. Constr. Mécan.* 1956, **88,** Apr., 367–371). The successful use of hot blast cupolas to provide hot metal for use in O.H. furnaces in Germany is discussed.—B. G. B.

New Canadian Steel Mill Instals British Melting Furnace. (*Brit. Engineering*, 1956, **39,** July, 28). Details are given of a three-phase direct-arc melting furnace made by G.W.B. Furnaces, Ltd., which is melting all the present all-scrap charges of Premier Steel Mills, Alberta. It has a nominal capacity of 10 tons, and rating of 4700 kVA, being fed by h.t. at 13,600 V. and 60 cycles. Output varies from 90 to 145 tons per day. The three 12-in. electrodes are controlled hydraulically by the Tagliaferra system.—K. E. J.

Report on Arc Furnace Future. F. E. Nicholson. (*Iron Steel Eng.*, 1956, **33,** April, 117–124). The author discusses the progress made in electric arc furnace design and operation in the last few years. The various aspects of these improvements are described under the headings of mechanical design, electrical equipment, fume collection, refractories, electrodes, and metallurgy.—M. D. J. B.

Experience in the Use of a Fettling Machine for Electric Furnaces. K. A. Mikhailov. (*Stal'*, 1956, (4), 360–361). [In Russian]. Brief descriptions are given of the design and operation of a machine for fettling electric steel-melting furnaces of about 40 tons capacity. The machine has been in successful use at the Chelyabinsk works for several years.

Small Melting Furnace Simulates Industrial Conditions. B. M. Larsen. (*J. Met.*, 1956, **8,** Aug., Section 1, 1057). Brief details are given of a small experimental furnace, heated by an electric arc and designed to reproduce the conditions of the basic open hearth and electric furnaces. Some of the applications of the furnace are mentioned.—G. F.

Electric Smelting Furnaces. M. J. Udy. (*J. Electrochem. Soc.*, 1956, **103,** June, 353–355). Data are presented on the operation of a small 100 kW electric smelting furnace. It is shown that for efficient operation, both the peripheral ohms factor and the kilowatt/in² of electrode, or electrode periphery in contact with slag must be considered. The application of this data to furnace design and electrode spacing is considered.—A. D. H.

When the Fire-bug was Present and the Steel Boiled Over Memories of the Beginning of a New Method of Steel Production at the Gerlafingen Works. H. Jenny. (*Roll'schen Werkzeitung*, 1956, **27,** May, 67–70). A description of the behaviour of the melt in an electric arc steel furnace.—U. E.

Behaviour of Oxygen and Nitrogen in the Electric-Furnace Bath with Oxidation by Oxygen or Ore. S. M. Gnuchev, G. K. Komissarov, and Z. V. Klochkova. (*Stal'*, 1956, (4), 323–327). [In Russian]. From full-scale investigations the following conclusions have been drawn: the oxygen content of the metal at the end of the oxidizing period (before removing the oxidizing slag) does not depend on the method of oxidation with oxygen or ore, the oxidizing period slags are less oxidized when the metal is blown with oxygen; the nitrogen content of the metal changes in the same way whichever oxidizing method is used; at the end of the reducing period (before deoxidation with aluminium) the oxygen content of the metal is 0·003–0·009% (for 0·07–0·42% C) for oxidation by oxygen or ore.—S. K.

Use of Oxygen in the Production of Steel in Electric Furnaces. Ya. M. Bokshitskii. (*Stal'*, 1955, (12), 1099–1104). [In Russian]. Soviet practice in the use of oxygen in electric arc steel melting furnaces is analysed, certain aspects being compared with American methods. The extension of oxygen utilization outside the low-carbon stainless steels field, especially to low-alloy tool and structural steel production is recommended. Among problems to be solved are achievement of improved desulphurization and decreased chromium oxidation, production of stainless steel using carbon ferrochromium in the charge, investigation of the advisability of stainless steel melting with a chromite furnace bottom, and establishment of conditions for melting down the charge using oxygen to secure desulphurization and reduce melting-down time by 20–30% for low-alloy steel production.—S. K.

Production from Induction Melting Furnaces. A. Tagliaferri. *Mét. Constr. Mécan.* 1956, **88,** June, 565–574). A comprehensive account of the design and productivity of induction furnaces is presented. Calculations showing the method of obtaining theoretical operating data are given.

Internal Stresses in Small and Medium Steelworks Moulds. J. Přibyl and Z. Kosňovský. (*Hutník*, 1956, **6,** (8), 239–243). [In Czech]. Internal stresses due to thermal gradients are analysed in terms of the heat transfer through moulds made of various types of cast irons. For the moulds considered a high pearlitic cast iron with 0·6–0·7% of combined carbon is recommended. The iron must be free from cementite, and the graphite, which should be lamellar, should be within the limits 2·75–2·95%.—P. F.

Effect of Melting and Teeming Technology of Rimming Steel on Flaws due to Lamination. P. S. Plekhanov, N. S. Mikhailets, A. E. Gorelkina and N. G. Nikulin. (*Stal'*, 1956, (5), 422–430). [In Russian]. Special experiments and the examination of routine data to determine the influence of melting conditions on the properties, particularly liability

to lamination, of rimming steel are described. After chemical composition the most important factor was found to be the tapping and teeming temperature. The influence of slag basicity before deoxidation or the FeO content of slag was not determined. Tendency to lamination was less when manganese loss during deoxidation was decreased. Optimal durations for boiling in the ingot mould were determined for two steels.

Influence of Low-frequency Vibrations of the Mould on Ingot Crystallization. V. I. Leont'ev. (*Problems of Metallography and Metal Physics, 4th Coll.* Moscow 1955, 70–76). Previous work has considered frequency and amplitude, the present investigations concerned range of vibration, velocity and acceleration. A ½ kg ingot was vibrated vertically at 20–50 kg with amplitudes of 0·1–1·0 mm and the microstructures examined after solidification. Preliminary experiments with transparent media were carried out to investigate mixing conditions and droplet expulsion, with air entrainment and formation of a zone containing bubbles. Experiments were also done on zinc with a closed mould and finally on steel. Conditions for grain refinement are outlined.

On the Inverted-V Segregated Zone of Large Carbon Steel Ingots. III. Segregation-Flaws Appearing in Large Carbon Steel Forgings. 3. M. Kawai. (*Tetsu to Hagane*, 1956, **42**, Apr., 301–306). [In Japanese]. The effects on flaw formation of casting temperature, Mn/Si ratio and type of annular segregation are investigated.—K. E. J.

In the Inverted-V Segregated Zone of Large Carbon Steel Ingots. IV. Defect in Segregated Line of Ingot and Segregation Flaw. M. Kawai. (*Tetsu to Hagane*, 1956, **42**, May, 383–391). [In Japanese]. The formation of segregation flaws in forgings is related to the non-metallic inclusions and cavities found at the boundary of the segregated zone in ingots, and to the type of annular segregation. Methods of reducing segregation flaws are described.—K. E. J.

Study on Blowholes in Iron and Steel. III. The Formation of Blowholes by CO Gas and Deoxidation. T. Fujii. (*Tetsu to Hagane*, 1956, **42**, Apr., 293–300). [In Japanese]. Blowhole formation was studied by a technique involving vacuum cutting and analysis of the gas. Blowholes in rimmed steel are produced by CO gas. In an experimental melt, the quantities of silicon, manganese, aluminium and oxygen were varied; limits of blowhole formation were found for various silicon contents, which varied with the carbon content, and for aluminium; manganese was too weak a deoxidizing agent to prevent blowholes. (13 references).—K. E. J.

Study on Blowholes in Iron and Steel. IV. The Formation of Blowholes by H_2 and N_2 Gas in Killed Steel. T. Fujii. (*Tetsu to Hagane*, 1956, **42**, May, 391–397). [In Japanese]. Hydrogen or nitrogen was lanced into molten steel, and the relation between blowhole formation and the amount of deoxidation by silicon or aluminium was established. In the case of deoxidation by aluminium, nitrogen gas was the main cause of blowholes.—K. E. J.

PRODUCTION OF FERROALLOYS

Tungsten Alloys. H. Willners and K. A. O. Ottander. (*Jernkontorets Legeringsmonografier*, 1956, No. 4, 37 pp.). [In Swedish]. A comprehensive survey is made of raw materials; manufacture of tungsten metal, ferrotungsten and tungsten melting base; World consumption statistics; price in Sweden; physical properties of tungsten, equilibrium diagram, segregation tendencies and thermal data; typical analyses, content of impurities and solubility of gases in tungsten; use in steel manufacture, recovery percentage and use for welding electrodes.—G. G. K.

Electric Smelting of Low-Grade Chromite Concentrates. J. P. Walsted. (*U.S. Bur. Mines. Rep. Invest.* 5268, 1956, Oct., pp. 28). Production of low-carbon ferrochrome is described.

Development of a Chloride Volatilization Process for Manganese Ores from Aroostook County, Maine: Progress Report. R. T. MacMillan and T. L. Turner. (*U.S. Bur. Mines, Rep. Invest.*, 5281, 1956, Oct., pp. 31). Ferromanganese can be produced.

Chloridization of Maine Manganese Ore. Preliminary Batch-Fluidization Tests on Maple Mountain–Hovey Mountain Samples. M. L. Skow, R. C. Kirby and J. E. Conley. (*U.S. Bur. Mines, Rep. Invest.* 5271, 1956, Oct., pp. 29). Ferromanganese is produced, though preferential hydrolysis of ferric chloride is possible.

FOUNDRY PRACTICE

Modern Cast Iron—A Review. M. Itzel. (*Gjuteriet*, 1956, **46**, (8), 106–110). [In Swedish]. After reviewing the mechanical characteristics of grey, malleable and nodular C.I., the author compares their properties, from the viewpoint of elongation, impact strength and machinability, with those of cast steel and stresses that a change from the latter to nodular C.I. gives a cost saving of 33%.—G. G. K.

The British Cast Iron Research Association. J. G. Pearce. (*Metallurgia*, 1956, **54**, Oct., 163–165). The activities of this association are described. Apart from the research programme the association has sponsored a number of very successful short courses and conferences. The development department which acts as an industrial advisory service together with an operational research team have been found of considerable benefit by the industry.—B. G. B.

The British Steel Castings Research Association. A. H. Sully. (*Metallurgia*, 1956, **54**, Oct., 178–182). Brief details are given of the activities of the association and of the construction of new research laboratories in Sheffield.—B. G. B.

Foundry Mechanisation with Fume Removal. (*Engineering*, 1956, **182**, Nov. 16, 638). A brief account of the Leyland Motors cylinder-block casting section.

Mechanised Foundry at Harwich, London Midland Region. (*Rail. Gaz.*, 1956, **105**, Dec. 7, 669–671). The British Railways foundry for the production of chairs, baseplates and brake-blocks is described and illustrated. Sand reconditioning is described.

Refractory Concrete for Foundry Floors. (*Canad. Metals*, 1956, **19**, Oct. 46). Refractory concrete slabs have been shown to resist the impact of molten steel at 158° C falling from a height of 7 feet.—B. G. B.

Discussion on the Hot Blast Cupola. (*Met. Ital.*, 1956, **48**, July, 237–240). [In Italian]. An account is given of a discussion on the advantages and disadvantages of using a hot blast in cupola working.—M. D. J. B.

Study on a Permanent Wall Type Furnace with [Imposed] Cooling. R. Sato and N. Tabata. (*Tetsu to Hagane*, 1956, **42**, May, 398–403). [In Japanese]. Results are quoted for the erosion of lining bricks in the cupola. Water-cooling could produce a permanent layer at the refractory wall, and tests were made on the melting zone. The method can greatly reduce erosion losses and maintenance costs, and give more uniform operation, with a higher proportion of scrap. Control of the cooling water is simple.—K. E. J.

On the Relation between Graphite Structure and its Spheroidisation in Cast Iron. Y. Masuko. (*Sumitomo Metals*, 1956, **8**, Apr., 53–64). [In Japanese]. Results are given for several types of iron melted in Tammann and high-frequency induction furnaces. Spheroidization is not clearly related to the content of minor " detrimental " elements or gases, e.g. Ti, V, As, Sb, Pb, Sn, O, N, or H; nor to the kind of pig-iron. It is markedly related to the graphite structure before treatment by cerium or magnesium; a flaky structure leads to easy spheroidization but the eutectic graphite structure does not. These structures may be mutually converted by pre-treatment, e.g. with Ca–Si, Fe–Si or hydrogen. The form of graphite is mainly influenced by the precipitating condition, particularly of the austenite phase, and for spheroidization solidification must occur in a critical temperature range. (27 references).

New Nodular Iron Process. H. K. Ihrig. (*Trans. Amer. Soc. Metals*, 1956, **49**, Reprint 7). A process is described in which as-cast nodular iron was prepared by using the halides of some alkali or alkaline-earth metals reduced by calcium-silicon. A mixture of sodium and magnesium chlorides was found to be the best. Chemical analyses, mechanical properties and photomicrographs are given.—E. E. W.

Reviews of the Main Characteristics of Spheroidal Cast Iron. D. Mensa and M. Noris. (*Ing. Mecc.*, 1956, **5**, June, 41–49; July 25–34). [In Italian]. After an historical review of the principal patents which have led to the production of spheroidal cast iron, the author discusses the various types of cast iron generally used in fields of engineering. The physical and mechanical properties of spheroidal iron are listed and discussed. Aspects of its workability are examined and case histories of spheroidal iron castings are given indicating the suitability of the material. (22 references).—M. D. J. B.

A Study of the Combustion of Coke in the Cupola. G. Ulmer. (*Fonderie*, 1956, June, 232–237). The aspects considered include the heat balance in the cupola, the air supply necessary for combustion, the temperature of combustion,

and the variations in the gas composition inside the cupola. In conclusion the effects of coke quality are considered.

Receivers, Tapping Systems and Slag Disposal. (*Found. Trade J.*, 1956, **101,** Sept., 27, 349–355; Oct. 4, 379–390). This report of Sub-committee T.S.44 of the Institute of British Foundrymen Technical Council is a comprehensive review of cupola tapping systems, receivers and slag disposal methods. Five different tapping systems are considered, particular attention being paid to the continuous front slagging system of which six variants are described. The advantages and disadvantages of continuous tapping are discussed together with the effects on metal temperature, and composition of continuous and intermittent tapping. Fixed, portable, heated and non-heated cupola metal receivers are described and experimental results on their relative efficiencies, and their effects on metal temperature, composition and structure are presented. The use of electric furnaces as metal receivers is discussed and their applications summarized. In conclusion slag containers and granulation methods of disposal are briefly described. (30 refs., bibliography).—B. C. W.

Rotary Furnaces: Batch Melting High-duty Cast Iron for Shell-moulding Process. A. J. Gibbs Smith. (*Iron Steel*, 1956, **29,** Sept., 437–438). The rotary furnace is well suited to melting high-duty cast iron for shell-moulding. Brief details are given of the design and layout of the Monometer type of rotary furnace, and of its lining life and fuel consumption.—G. F.

Precision Melting. (*Canad. Metals*, 1956, **19,** Aug., 52–54). A description is given of a modern rotary furnace for iron melting with a capacity of one ton.—B. G. B.

Outline of Recommended Melting Practices for Small Heats. N. J. Grant. (*Indust. Heating*, 1956, **23,** May, 968–978, 1112). The influence of crucible material, type of furnace, alloy composition, hydrogen pick-up, deoxidation practice and temperature measurement and control on the quality of 5–20 lb. melts for investment castings are considered. Details are given of recommended practice for type 321 stainless steel.—A. D. H.

The Influence of Casting Temperature on the Grain Size of Grey Cast Irons. M. Ferry and J.-C. Margerie. (*Fonderie*, 1956, June, 223–231). The factors affecting the grain size (i.e. the size of the eutectic colonies) of grey cast iron are discussed and the results of experiments on the influence of casting shape, casting size, and pouring temperature are presented. In general, decreasing casting size and low pouring temperatures gave a fine grain size, although pouring temperature had little effect on the grain size of low silicon irons. The grain size at the surface of geometrically similar castings was found to depend on the parameter $\frac{V.\Delta T.}{S}$ where V and S are the volume and surface area respectively of the casting, and ΔT is the difference between the pouring temperature and the eutectic temperature. Differently shaped castings gave different curves for grain size as $\frac{V.\Delta T.}{S}$. The grain size at the centre of the casting was found to be almost independent of the casting shape and to vary principally with the parameter $\frac{K.V}{S}$ where K depends on the pouring temperature.—B. C. W.

Ladle Additions to Grey Cast Iron for High Duty Parts. Without Disturbance of Normal Production. C. R. van-der Ben. (*Alloy Metals Review*, 1956, **8,** Sept., 2–6). Products to meet BSS 1452 are produced for special jobs by addition of Ni or Mo to the ladle. Microstructures and typical parts so made are illustrated.

A Note on the Subject of the Variation of the Mechanical Properties of Grey Cast Irons with Thickness. M. Ferry. (*Fonderie*, 1956, July, 267–273). The variation of the tensile strength (R), elastic modulus (E) and hardness (H) of grey cast iron with the diameter of the test casting is discussed and experimental results on a number of cast irons are analysed. The results are expressed in terms of the relationship between $\log R/R_0$, $\log E/E_0$ and $\log H/H_0$, and between R and $\log \phi$, where R_0, E_0 and H_0 are the values obtained with bars of 30 mm diameter, and ϕ is the test bar diameter. The dependence of these relationships on the chemical composition of the iron is discussed.—B. C. W.

The Effect of Isothermal Heat Treatment on the Wear Characteristics of Cast Iron. J. Czikel and H. Münch. (*Giessereitechn.*, 1956, **2,** Aug., 177). The wear characteristics of castings were improved by intermediate stage heat treatment. Cast iron specimen cylinders were placed in a rotating drum and the detritus was measured. The best temperature range for the heat treatment was found to be from 280 to 320° C. Brinell hardness will not rise above 350 and the wear resistance can be increased by a factor of 3.—L. J. L.

The Effect of Phosphorus on the Foundry Properties and the Structure of Grey Cast Irons. M. Ferry. (*Fonderie*, 1956, Feb., 55–67). Previous work on the effect of phosphorus on the castability and soundness of grey cast iron is reviewed. The results of some experiments to investigate the effects of phosphorus and casting temperature on the soundness of four different types of castings made with two different cast irons (high and low silicon respectively) are then presented. The effects of increasing phosphorus content on the structure of the two base irons was also studied with particular reference to the distribution of the phosphorus eutectic and the segregation associated with its solidification.—B. C. W.

Influence of Boron on Tensile Strength and Structure of Malleable Iron. G. Östberg. (*Gjuteriet*, 1956, **46,** May, 72–76). [In Swedish]. Tensile tests to establish the influence of boron on steel properties have confirmed that tensile strength, yield-point and elongation are reduced by this element. It was observed that malleablizing was delayed in the high temperature range and accelerated at low temperatures. Microscopical studies did not indicate any influence of boron on the primary structure of white cast iron, although the shape of the manganese sulphide particles was changed.

Corrosion-resistant High Silicon Castings. (*Aciers Fins Spéc. Franç.*, 1956, March, 79–82). A number of ferro-silicon alloys have been developed which are extremely resistant to corrosion. Details are given of their characteristics; with a brief note on points to be considered in the design of parts when using such alloys.—E. A. C.

Steel Castings for Earth Moving Equipment and Agricultural Machinery. (*BSFA pamphlet*, 1956, pp. 15).

Take a Fresh Look at Foundry Moulding Sands. J. M. Anspach. (*Foundry*, 1956, **84,** Aug., 100–103). The fundamental factors affecting the choice of moulding sands are discussed with particular reference to the importance of grain size and clay bond characteristics.—B. C. W.

Sand Control Important in Production of Chemical Castings. E. Bremer. (*Foundry*, 1956, **84,** Sept., 129–133). A description is given of the moulding, sand handling and sand preparation methods used at the Duriron Co. Inc., Dayton, O., in the production of high-silicon iron castings for the chemical industry.—B. C. W.

You Can Improve Your Sand. E. E. Woodliff. (*Foundry*, 1956, **84,** Oct., 122–125). The author stresses the importance of sand control and discusses the practical aspects of a control system. The defects in grey cast iron which are associated with the condition of the sand are then briefly considered and recommendations are made to avoid their occurrence.

Automatic Mix for Moulding Sand. (*Steel*, 1956, **139,** July 16, 145).

Steel Penetration—Some Further Observations. R. C. Emmons, J. B. Jones, and J. R. Bach. (*Foundry*, 1956, **84,** Oct., 150, 153). The results of a further experiment on the mechanism of steel penetration into sand moulds are discussed in relation to previous work. Penetration is now attributed to the effects of moisture firstly in opening up the sand by removing silica from the mould face, and secondly in breaking down the iron carbonyl in the cooler parts of the sand. The resistance of green sand to steel penetration is explained by the effect of moisture in preventing the formation of the carbonyl.—B. C. W.

Bright Future Seen for Vacuum Investment Castings. F. K. Iverson. (*Iron Age*, 1956, **178,** July 19, 120–122). There are indications that vacuum investment castings made from vacuum melted alloys may have very desirable properties. Comparison is made between air, argon, and vacuum investment casting, including an example of cost figures. A low-cost vacuum furnace is desirable.—D. L. C. P.

Pattern Dies for Investment Casting. D. E. Brooks. (*Iron Steel*, 1956, **29,** July, 349–352). The author describes the types of pattern die used in investment casting and discusses the factors and fundamental considerations influencing the die design. The techniques of making pattern dies are outlined.—G. F.

The Determination of the Pouring Time of Castings. C. Trenclé. (*Fonderie*, 1956, July, 251–266). Five conditions

that fix the upper and lower limits for the pouring time of a casting are discussed in detail. The conditions are that (1) solidification does not begin until the mould is filled, (2) none of the risers solidify during pouring, (3) scabbing and rat-tail defects are avoided, (4) erosion of the mould and displacement of the cores does not occur, and (5) adequate time is allowed for venting of the mould. Conditions (2) and (3) are considered to be the most important and the practical conclusions arising from the discussion are then considered. An equation from which the pouring time for a given casting can be determined is presented, and graphs are drawn for cast irons poured at 1300° C from which the pouring times of differently shaped castings made in various types of moulds may be obtained. (26 refs.).—B. C. W.

The Risering of Nodular Iron. R. C. Shnay and S. L. Gertsman. (*Canad. Metals* 1956, **19,** Aug., 48–51; Sept., 40–50). The influence of change in pouring temperature on the shrink depth of nodular cast iron has been studied. The effect of mould rigidity was also investigated. Cooling rate curves and metallographic studies were made to determine the causes of the influence of pouring temperature. The results obtained stress the importance of using the metal at the correct temperature for casting. (14 references).—B. G. B.

Soluble Cores. W. A. Dubovick. (*Precision Met. Mold.* 1956, **14,** July, 38–40). The use of water-soluble wax cores in the production of complex investment castings is described.

Factors Involved in Blowing Shell Moulds and Cores. G. A. Conger. (*Foundry*, 1956, **84,** Apr., 154–155, 158, 160, 162, 164, 167, 168). The factors discussed include the cold and hot coating of sand, and the design and operation of the equipment for blowing moulds and cores. The production of hollow shell cores, and the use of permanent mould liners are briefly described.—B. C. W.

The Croning Method in the Foundry of the Volkswagen Works. G. Serwe. (*Giesserei*, 1956, **43,** May 24, 285–287). The use of the Croning method in the Volkswagen foundry for the manufacture of cylinders and cylinder head cores is described and illustrated by 8 photographs. The large number of pieces necessitates a well planned installation. The finishing of the cylinders is also described.—R. J. W.

CO_2 Process: Formation and Dissolution Mechanism in Relation to the Silicate Bond. C. Starr. (*Found. Trade J.*, 1956, **101,** Sept. 6, 265–266). The reactions of CO_2-bonded silicate cores under various temperature conditions are described and the reasons for the good breakdown with high and low melting-point alloys and the poor breakdown with medium melting-point alloys are explained. The requirements of any addition to improve breakdown are discussed.

Indian Foundry Uses CO_2 Process to Advantage. C. A. Phalnikar. (*Foundry*, 1956, **84,** Aug., 140–142, 144, 146). Experiences with the use of the CO_2 process in the grey iron foundry of Kirloskar Brothers Ltd., Kirloskarvadi, India, are described.—B. C. W.

How to Handle Carbon Dioxide. W. Bigelow, jun. (*Foundry*, 1956, **84,** Sept., 203, 205, 207). Instructions for handling and using cylinders of liquid CO_2 in the foundry are summarized.—B. C. W.

Chambersburg Engineering Foundry Specializes in Large Precision Castings. F. Mueller. (*Steel Processing*, 1956, **42,** July, 393–395, 414). A general description of this new foundry is given with a detailed layout. Castings up to 150,000 lbs are made; blast furnace blower cases and hydraulic press cylinders are typical examples. One cupola of 45,000 lbs/h melting capacity is used.—A. H. M.

Sulzer Precision Castings. W. Sulzer. *Sulzer Tech. Rev.*, 1956, **38,** (1), 1–13). An illustrated account of the methods used with a discussion of the collaboration necessary between the designer and foundryman.

Production Increased by New Stack Moulding Technique. K. L. Mountain. (*Foundry*, 1956, **84,** Sept., 236, 239). A simplified stack moulding technique is briefly described. It is claimed that the technique almost doubles mould production whilst reducing sand requirements by about 40% and casting costs by about 20%.—B. C. W.

Gauge Control for Investment Castings. J. E. Ingalls. (*Foundry*, 1956, **84,** Sept., 162–166, 168). The need for a control system for gauges when large numbers of complex castings are being produced to accurate dimensions, is stressed. Recommended inspection and maintenance methods are then discussed in detail.—B. C. W.

Economics of Investment Cast Vacuum Alloys. F. K. Iverson. (*Prec. Met. Mold.* 1956, **14,** June, 130–135). The argon and vacuum investment casting processes are briefly described and their economics compared with the standard air melting and casting process.—D. H.

Quality Control of Investment Castings. Part 1. C. Yaker. (*Precision Met. Mold.*, 1956, **14,** Jan., 47, 79–80). The author lists and discusses eight control requirements which must be agreed upon by producer and consumer in order to produce investment castings of aircraft quality.—D. H.

When Should you Use Investment Castings? W. O. Sweeny and D. M. Stagg. (*Precision Met. Mold.*, 1956, **14,** June, 44–45). A simple guide on which the decision to use investment castings may be made is given.—D. H.

Practical Approach to Quality Control in the Gray Iron Foundry. C. F. Walton. (*Bull. School Mines Met. Univ. Missouri*, 1956, No. 91, 24–34). Various factors of the need for uniformity are discussed and methods of securing it. In particular, analyses of samples during furnace heats are given showing the large variations that may be found.

Quality Control of Cast Iron and Steel Melts Using Optical Measurement of Temperature. K. Orths. (*Arch. Eisenhüttenwesen*, 1956, **27,** May, 289–295). The use of radiation analysis, and the effects of mould material, composition, pouring temperature, and piece weight on the solidification of iron and centrifugally cast piping are indicated. The optical method of thermal analysis is described in its application to centrifugally cast piping. Reduction temperatures and equilibria, colour pyrometry, and the limits of overheating are discussed.

Effect of Technological Factors on the Design of a Centrifugal Tube-Casting Machine. A. A. Mikhel'son (*Stal'*, 1956, (4), 348–351). [In Russian]. Full-scale tests on centrifugal tube-casting machines have been carried out to investigate a number of technological process parameters. Improvements in the design of various parts of such machines have been effected.—S. K.

Some Fundamental Studies on the Shell Moulding Process. T. Mishima, Y. Mishima, and K. Kobayashi. (*Tokyo Univ. Fac. Eng. Annual Report*, 1955, 109–113). The process is reviewed and attention directed to selection and control of sand, optimum height of dump box and comparison of stripping agents. Silicon esters and silicones were used as stripping agents.

Rotary Furnaces for Shell Moulding. (*Brit. Engineering*, 1956, **39,** Aug., 49). Modern rotary furnaces melt to consistent analysis, to rigid times, and with metal temperatures which may exceed those of the cupola. They have economic advantages. Those for iron melting have cold-charge capacities of 0·5–5·0 tons. Details are given of a 1-ton Monometer furnace installation; on continuous melting cycles, 1 ton of high-duty iron at 1500° C is produced in 1 hour, with a fuel consumption of 25–30 gal/h and refractory usage of 50 lb/ton.—K. E. J.

Casting Camshafts in Shell Moulds. K. L. Mountain. (*Foundry*, 1956, **84,** Sept., 124–128). The operation of a mechanized shell moulding production line for camshafts is described. The line includes three automatically controlled machines which blow the shell moulds on heated patterns and then cure them under a pressure of about 300 lb/in². These blow-hot press machines are claimed to have more than doubled the production of camshafts.—B. C. W.

Electric Motor Parts Cast in Shell Moulds. J. B. Shallenberger. (*Foundry*, 1956, **84,** Sept., 276, 278). The patterns, coreboxes and machines used in producing electric motor parts by shell moulding are very briefly described.—B. C. W.

The Development of Shell-Moulding Methods in the British Commonwealth. J. Fallows. (*Giesserei*, 1956, **43,** May 24, 287–291). A review article in which the development of the Croning methods of shell moulding in Great Britain are followed. Economic aspects are discussed. Future possible developments are considered. Development of the method in the British Empire is also described.—R. J. W.

Pipes, Depressions and Cavities in Iron Castings. J. Pascal. (*Mét. Constr. Mécan.* 1956, **88,** Mar., 249–257; Apr., 347–357). The fundamental principles controlling the method of solidification of iron is discussed with special reference to the formation of defects in iron castings.—B. G. B.

A Study of the Cooling Process of Large Iron Castings in Sand Moulds. J. P. Jegorenkow. (*Giessereitechn.*, 1956, **2,** Aug., 173). Results of temperature measurements on large

castings by means of thermocouples are reported. The cooling process can be divided into 5 stages. In the first, the heat of superheating is dissipated; freezing occurs in the second; further cooling, to the conversion of austenite, in the third; gamma–alpha conversion occurs in the fourth, and cooling to room temperature in the last stage. In order to keep stresses to a minimum crack-resistant castings should be knocked out at 700–750° C.—L. J. L.

Development and Application of a Foundry Test for Assessing Shrinkage Behaviour in Grey Cast Iron. J. Grice and J. T. Berry. (*Found. Trade J.*, 1956, **101,** Aug. 16, 159–170). Previous work on metal shrinkage is reviewed with particular reference to Chvorinov's rule and the work of Pellini, and results obtained in an investigation into the shrinkage behaviour of grey cast iron are reported and discussed. It was found that to avoid unsoundness in the casting the minimum volume/surface area ratio for a cylindrical feeder on a 6 in. cube casting was 1·0 with a sand feeder, 1·1 with exothermic cover and 0·725 with an exothermic-sleeve feeder. The cube test casting had several disadvantages and subsequent work on the efficiency of feeders was done with a horizontal long cylinder fed and gated at the centre. The time-temperature changes along the bar during solidification and examination of the castings for shrinkage defects showed that Chvorinov's rule is only partially successful, and that the criterion of Pellini correlating temperature plateaux during solidification and centre-line shrinkage is not entirely applicable to grey cast iron.—B. C. W.

Results of Experiments on Vacuum Casting of Steel. Part I. The Molten State. Z. Eminger and F. Kinský. (*Hutnické Listy*, 1956, **11,** (6), 345–355). [In Czech]. Scope and methods of vacuum casting are reviewed. In the present paper, the first of a series of three, equipment used by the authors in the experimental study of vacuum casting of ingots of plain and Cr–Ni–Mo steels weighing up to five tons are described and the results are discussed with reference to micrographs, chemical and gas analysis on ingot sections, grain size determinations, flaking, non-metallic inclusions, formation of blow-holes, and the tendency to overheating. Shrinkage, inclusion content and hydrogen content of the steels were reduced in all cases. The vacuum treatment appeared to have little effect on the distribution of impurities and the impact strengths. The vacuum cast steels have a finer austenite grain, are less prone to overheating than similar steels cast by usual methods, and the tendency to flaking in Cr–Ni–Mo steels was greatly reduced or entirely eliminated.

The Effects of Calcium Hydride Addition to Grey Cast Iron. T. Kusakawa and A. Kobayashi. (*Tetsu to Hagane*, 1956, **42,** Apr., 338–343). [In Japanese]. Experiments with Swedish charcoal pig-iron showed that addition of CaH_2 to iron covered by a flux of fluorspar produced spheroidal graphite cast iron, although some quasi-flake graphite was always present if no other additions were made. The additions produced blow-holes, which disappeared when the spheroidal structure appeared, and desulphurization was effected. Addition of 2% CaH_2 did not reduce the amount of magnesium required for the spheroidal structure.—K. E. J.

The Improvement in Properties of Molten Iron. III. The Effect of Mill Scale Addition in Comparison with Oxygen Blowing into Molten Iron. T. Tottori. (*Tetsu to Hagane*, 1956, **42,** May, 379–383). [In Japanese]. Scale was added at about 1450° C to minimize oxidation effects. The amount of titanium and vanadium in the iron decreased, and hence irons normally unsuitable for spheroidization could be so converted by magnesium treatment. The effects of oxygen blowing, which produced a great increase in temperature, were similar but more marked.—K. E. J.

Production of Heavy Steel Castings. S. Taylor. (*E.S.C. News*, 1956, **10** (3), 4–9). An illustrated account of work at Grimesthorpe Foundry.

Moulding Characteristics of Jubbalpore Sands. R. M. Krishnan and B. R. Nijhawan. (*J. Sci. Indust. Res.*, 1956, **15A,** July, 314–327). Green, yellow and white sands were subjected to various tests including green compressive strength, mould hardness, A.F.A. and casting characteristics. Green and yellow sands proved suitable for steel foundry purposes but white gave poor results.—E. E. W.

Procedures for Risering Steel Castings. H. F. Bishop and W. H. Johnson. (*PB* 111871, 1955, pp. 19; *TIDU List* 578). Rules drawn up from earlier reports are applied using formulæ (1) for user feeding range in uniform sections, (2) for feeding range in heavy and light parent-appendage combinations and (3) for the application of chills to increase riser feeding range. Calculation for minimum riser dimensions are shown.

CO_2 Speeds Coremaking. (*Steel*, 1956, **139,** Aug. 6, 111–114). The production and use of carbon dioxide-hardened cores and moulds is described. The advantages are leading to its increasing adoption in U.S. foundries of many types.—D. L. C. P.

A Practical Guide to the Design of Grey Iron Castings for Engineering Purposes. (*The Council of Ironfoundry Associations Publications*, 1956, pp. 55). A translation of a publication of the Commission Techn. de la Métallurgie des Fontes, Oct., 1954. Properties and structures, rules for design, gating problems, venting methods, mould joints and taper, coring and fettling are considered with an appendix on pattern making.

Large Hammer-block Casting for Axle Forge. (*Metal Treatment*, 1956, **23,** Sept., 366). This 143-ton block made by English Steel Castings Corporation replaces an 82-ton two-piece block previously used in the axle forge of Steel, Peech and Tozer. Arrangements for delivery and installation are outlined.—A. H. M.

Extending Applications of Continually Cast Iron Bar. (*Machinery*, 1956, **88,** May 25, 848–849). Uses of the Flocast bar of the Harold Andrews Sheepbridge Co. are mentioned and parts illustrated. Machining conditions are recommended.

New Advances Swell Shell Molding Uses. O. W. Winter. (*Iron Age*, 1956, **178,** Oct. 11, 108–110). **Improved Mold Sands drop Shell Casting Costs.** (Oct. 18, 132–134). A two-part article on the method reviewing advantages, costs, sands, additives, and cores.

Increasing Productivity by Shell Moulding. (*Machinery*, 1956, **89,** Sept. 21, 678–684). Tappet guide making by British Piston Ring Co. is described in detail. A mould of 8 components weighs 4 lb 10 oz whereas 15 were previously made from a 14 lb bar.

Metallurgy and Technology of Investment Casting. K. A. Krekeler. (*Stahl u. Eisen*, 1956, **76,** Oct. 18, 1360–1365). The lost-wax precision investment casting process and its applications are reviewed in detail. Compositions of ferrous and non-ferrous alloys suitable for this casting process are given with particular emphasis on heat-resistant alloys for turbo-jet engine blading. Other applications of the process are quoted, mainly for tool tips and magnets.—T. G.

Solidification Rate of White Cast Iron. Y. Masuko. (*Tetsu to Hagane*, 1956, **42,** Apr., 306–313). [In Japanese]. A linear relation between the thickness of solidified outer layer and the square root of solidification time is found only in the early stage of solidification. The constant of the equation (rate of skin formation) decreased as the casting temperature, mould temperature or casting size increased. The full solidification curve is **S**-shaped, with two inflections which are associated with the beginning of crystallization of a "bee-hive"-like ledeburite and the end of crystallization of dendritic cementite.

Suppression of Microporosity in Defective Castings. L. Gascuel. (*PACT*, 1956, **10,** Aug., 331–332). An autoclave method of impregnation of porous sections is described. A thermosetting resin is used.

Dust Measurement for Control of Foundry Atmospheres. A. Ahlmark and H. Ohman. (*Gjuteriet*, 1956, **46,** (8), 101–106). [In Swedish]. Details are given of Swedish investigations into the incidence of silicosis in foundry workers and of the relationship between dust concentration and illness rates. The results of dust measurements in various sections of ordinary and mechanized foundries are presented and the importance of correct sampling is stressed.—G. G. K.

Ventilation and Dust Collection in Melting Bays. V. Almborg. (*Gjuteriet*, 1956, **46,** (6–7), 86–91). [In Swedish]. Illustrated examples are given of the use of metallic screening, gas dilution, local extraction, wall ventilators, smoke doors, chimney stack ejectors, hoods, and wet and dry dust separators in order to remove the smoke and dust encountered with crucible furnaces, electric furnaces and cupolas.—G. G. K.

Foundry Ventilation from the Viewpoint of Worker Protection. E. Pelow. (*Gjuteriet*, 1956, **46,** (6–7), 83–85). [In Swedish]. After citing relevant sections of the Swedish Factory Acts concerning the provision of ventilation, the author describes investigations made since 1911 by various bodies with the aim of improving working conditions in foundries, with special reference to heat and dust. Some hints are given on the use of measures other than ventilation for reducing gas and dust nuisances.—G. G. K.

VACUUM METALLURGY

Problems of Vacuum Technology in Metallurgy. K. Diels. (*Berg Hüttenmann Monatsh.*, 1955, **100,** [7–8], 201–207]. Furnaces are described for degassing steel and other metals, and degassing during pouring is discussed. Inclusions are reduced and ghost lines are eliminated. Sintering is then considered and heating arrangements briefly described. Melting and alloying processes are then discussed and the apparatus used outlined. Sublimation, distillation and evaporation, the last-named as a coating process, are finally dealt with.

Recent Developments in Vacuum Heat-treating Furnaces. R. R. Giler. (*Indust. Heating*, 1956, **23,** Oct. 2038–2050). Vacuum-annealing furnaces, chiefly the Kold-Retort furnace, are described, principally for treating titanium.

Vacuum Melting: Large Range of Furnaces Now Made in England. (*Iron Steel*, 1956, **29,** Nov., 523–524). Efco-Edwards Vacuum Metallurgy Ltd. are now producing a range of five vacuum-melting furnaces, with capacities of ½ lb, 28 lb, 56 lb, 5 cwt, and 1 ton. Details are given of the design and performance of each type.—G. F.

On the NRC 50 kg Commercial Vacuum Melting Furnace. T. Aoyagi. (*Sumitomo Metals*, 1956, **8,** Apr., 109–116). [In Japanese]. After an introduction to the principles of vacuum melting, the construction and operating characteristics of the National Research Corporation furnace are described.

REHEATING FURNACES AND SOAKING PITS

Study of the Influence of the Furnace Atmosphere on the Scaling of Ingots. J. Moreau. (*Centre Doc. Sidér. Circ. Inform. Tech.* 1956, No. 10, 1973–1982). The results are given of a study of the formation of scale on a carbon steel heated at high temperatures in different furnace atmospheres at 10 plants in France. The furnaces were gas, oil or coke fired and the temperatures ranged from 1050 to 1300° C. The oxygen content of the exit gas varied between 0 and 15%. Photomicrographs of the surface structure of the scale are shown and the structure obtained are discussed in detail.—B. G. B.

Heating and Scale. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1956, No. 10, 1969–1972). A summary is given of recent research work by IRSID on the structure of the scale on ingots before and after passage through the soaking pits. Attempts to reproduce scale under controlled laboratory conditions are described.—B. G. B.

Rapid Heating of Small Forgings. J. E. Russell. (*Metal Treatment*, 1956, **23,** July, 251–256). The author outlines the advantages of rapid heating and reports various results. Methods of heating are dealt with, and include furnace induction and electric-resistance heating.—A. H. M.

Experience with Atomized-Oil Burners in a Pusher-Type Billet Reheating Furnace. E. Neumann and J. Seitz. (*Stahl u. Eisen*, 1956, **76,** Oct. 18, 1356–1360). Ideal combustion conditions and metallurgical requirements in reheating furnaces are often incompatible. Choice and arrangement of burners is more important in oil-fired furnaces than in gas-fired furnaces. The authors report experiments with oil burners using preheated air (400° C). The burners contain a turbo-atomizer. The oil heated to more than 100° C enters the atomizer via a tangential slot—the length of which is variable for controlling the oil consumption—and leaves the chamber in the form of an atomized jet through a nozzle. Perfect atomization and instantaneous combustion is essential in order to avoid oil droplets falling on the billets thus carburizing the steel surface. The preheated air passes a turbine for increasing the turbulence and better mixing with the oil jet. The oil consumption can be varied between 25 and 310 kg per h by using the slot control, different orifices for the oil jet, or oil pressures between 2 and 13 atm.—T. G.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

The Heat Treatment of Steel. C. A. Siebert. (*Metal Treating*, 1956, **7,** Sept.-Oct., 14–16, 42). Isothermal transformations and the method of constructing the T-T-T diagram are discussed. The behaviour of steel during quenching is presented in terms of the above considerations.—A. H. M.

The Heat Treatment of Forgings. A. O. Schaefer. (*Metal Treating*, 1953, **7,** Sept.-Oct., 6–8, 20–22, 24). Some aspects of heat treatment which are peculiar to forgings are discussed. These are: the avoidance of flakes, quenching techniques of "critical sections," and shape stability of forgings.—A. H. M.

The Distortion and Dimensional Instability of Heat-treated Components. W. B. Kemmish. (*Machinery*, 1956, **88,** May 18, 759–764). Factors in the retention of shape and dimensions is heat-treatment and in service are discussed.

Mechanisation and Automation in the Heat-treatment of Metals. H. Diergarten. (*Härtereï-Tech. Mitt.*, 1956, **12,** 9–23). A critical survey of methods and means for automation in the heat-treatment of metals is given. It is concluded that mechanization and automation can be improved, resulting in lower cost, less manual work and in a better product with improved reproducibility.—R. P.

Alloy Salt Pots of Welded Construction Relieve Heating Problems. M. H. Minks. (*Indust. Heating*, 1956, **23,** Oct., 2126–2130, 2138). Neutral salt bath furnace pots of Inconel are satisfactory for operation at 1500–1600° F. Construction is described.

Useful Electric Furnaces. (*Machinery Lloyd*, 1956, **28,** Nov. 17, 53–54). A furnace developed for high-speed steel hardening and quenching out of contact with air, and another with mesh belt conveyor for bright annealing and brazing are described.

Electrical Controls make Industrial, Gas-Fired, Special Atmosphere Furnaces Safe. C. E. Huggins. (*Gen. Motors Eng. J.*, 1956, **3,** Oct.-Nov.-Dec., 9–11). Circuits devised at Delco Products Division are shown and their advantages outlined.

Efficient Instrumentation for Oil or Gas Fired Heat-Treatment Furnaces. L. Walter. (*Metal Treatment*, 1956, **23,** Oct., 420–422). Various types of oil and gas fired heat-treatment furnaces are considered in relation to suitable instrumentation for control. Relative problems of equipping existing and new plant are discussed, and the desirable characteristics of effective systems of instrumentation outlined.—A. H. M.

Radiation Cataract. (*Metal Treatment*, 1956, **23,** Sept., 358). The article describes an investigation at the rolling mills of the Butterley Co. Ltd. into the effects of intense radiation from hot metal on the eyes of workers. Ophthalmological examination and radiation measurements were conducted. A hypothesis of individual susceptibility is favoured, and it is concluded that radiation cataract might develop as a result of normal every-day life radiation.—A. H. M.

Hard and Extra-hard Monophasic Coatings obtained by Diffusion. H. Buckle and J. Moisan. (*6th Internat. Mech. Eng. Congress*, 1956, reprint A1/13, pp. 9). Mainly non-ferrous.

Application of Induction Heating to Metallic Diffusion. N. A. Lockington. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C1/5, pp. 12). Si and Mn in mild steel and Al and Cr in Ti are outlined. Diffusion into Ni and Mo are mentioned.

Longer Service Life through Hard Surfacing for Petroleum Industry Equipment. H. S. Gonser. (*Welding J.*, 1956, **35,** Sept., 890–894).

Gas Carburizing at the Ford Motor Co. Ltd. (*Wiggin Nickel Alloys*, 1956, (42), 2–3). The furnace is described and a plan and illustrations given.

Practical Aspects and Methods of Carbon Control in Heat Treatment. W. T. Groves. (*Indust. Heating*, 1956, **23,** Oct., 2060–2072, 2250–2254). Hardnesses and residual stresses in carburized bars are shown and control of atmospheres and carbon oxide equilibria are considered. Dewpoint effects are discussed.

Surface Effects in the Heat-treatment of Die Casting Dies. F. A. A. Crane. (*Machinery*, 1956, **88,** Apr. 27, 576–578). Carburization and decarburization and their effects are considered and the use of packing media.

The Increase of Tensile Strength in Structural Steel by Localised Heat Treatment. H. Staudinger. (*Schweiz. Arch. Wiss. Tech.*, 1956, **22,** Aug., 244–257). The author shows that the endurance of structural steel may be improved by localised heat treatment if the internal stresses consequent on the treatment are in such a direction and of such a magnitude that they oppose the stress applied in use. The technique is especially suitable at grooves, fillets, holes and cross-sectional tapers in members and may also be applied to heavy duty parts in internal combustion engines such as crank-shafts, connecting rods and springs.—J. R. P.

Finish Tiny Stainless Steel Parts, Then Heat Treat. (*Iron Age.*, 1956, **178,** Aug. 16, 95). A special heat treatment is

described, before which, small stainless steel parts can be finish machined. The hardening or toughening treatment does not cause discoloration or distortion. Careful control of furnace atmosphere is necessary, and vacuum treatment is used for tempering and annealing.—D. L. C. P.

Continuous Gas-carburizing of Steel Components. T. W. Ruffle. (*Metal Treatment*, 1956, **23,** Sept., 339–342). The layout and a description of a new Birlec furnace are given. The steel components handled are mainly gears and automotive parts. Cycle of operations, metallurgical control and operational data are discussed.—A. H. M.

On the Case-hardening of Steels by the Modified Gas Utilising Oxygen. III. M. Okamoto and N. Shirai. (*Tetsu to Hagane*, 1956, **42,** May, 421–425). [In Japanese]. Steels were heated in a closed vessel, and reactions between the cyanide salt and oxidizing agents (air, water or Fe_2O_3) were studied. The theoretical optimum ratio gave the maximum hardness to the steels.—K. E. J.

Cheaper Case for Stainless. (*Steel*, 1956, **139,** Aug. 6, 115). Case hardening of stainless steels by liquid salt bath is described. It is simpler and less expensive than gas nitriding. The case is thinner; hardness tests are reported for several steels.—D. L. C. P.

Assumptions in Case-Hardening for Obtaining Optimal Properties in Parts. R. Fizia. (*Härterei-Tech. Mitt.*, 1956, **10,** 29–36). This paper deals with conditions for obtaining optimal tensile strength and complete martensite formation. The advantages of hardening from austenite are discussed. Applications of the proposed methods are suggested.—R. P.

Present State of Flame Hardening Processes. H. W. Groenegress. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/6, pp. 10). Flame hardening of gears, crankshafts, wheels and rails is outlined and the replacement of acetylene by town's gas, use of the "milliscope" and choice of steel discussed. Cast iron hardening is mentioned and economics of the processes considered.

Surface Hardening by Induction Heating for Increasing Wear Resistance. O. Liedholm. (*6th Internat. Mech. Eng. Congress*, 1956, reprint A1/9, pp. 17). The principle is outlined and depths of penetration indicated. Choice of apparatus is discussed and the importance of correct quenching noted. Practical methods are then described and results obtained.

High-Frequency Induction Surface Hardening and its Uses in Mass Production in European Industry. Labrousse. (*6th Internat. Mech. Eng. Congress*, 1956, preprint, A1/8, pp. 9). A general review with an example of contour hardening is given.

Optimum Specifications for an Induction Heating Plant for Surface Hardening of Steel Parts. Descarsin. (*6th Internat. Mech. Eng. Congress*, 1956, preprint, Section A1/3, pp. 16). Frequency should correspond to depth of penetration required, power should allow of nearly instantaneous heating and quenching should be immediate. Two generators of different depths of penetration might be used. A high-frequency generator is recommended for general purposes if only one is to be installed.

Induction Heat Treating. T. A. Dickinson. (*Metal Treating*, 1956, **7,** July-Aug., 6–7). This note describes the induction units used by a heat treating firm, and outlines their experience in this field.—A. H. M.

Induction Hardening Operations on Crankshafts. (*Machinery*, 1956, **89,** Nov. 16, 1122–1125). An Elotherm plant is described.

Control Problems in Induction Surface Hardening. C. di Pieri. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/4, pp. 13). Circuits for h.f. vacuum-tube generators are discussed.

Induction Hardening as a Means of Increasing the Resistance to Wear of Internal Combustion Engines and Machinery. G. W. Seulen. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/11, pp. 11). Apparatus and effects are described.

Surface Hardening of Machine Parts by the Electrospark Process. L. A. Mirkin. (*Vestnik Mashinostroeniya*, 1955, **35,** (4), 48–51). Gearbox components were used and hardening and testing methods are described. Wear resistance and fatigue limit were measured and surface stresses examined. Capacitances and choice of electrodes are discussed and the only disadvantage recorded is the low rate of production.

Gas Carburizing and Carbonitriding. L. G. W. Palethorpe. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/7, pp. 20). The furnace and furnace atmospheres, results obtained and uses are reviewed.

Dynamic Characteristics of Steels after Surface Hardening by Carbonitriding. D. Giaccone and R. Battini. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/5, pp. 14). It is found that NH_3 content does not greatly effect fatigue properties unless after water quenching which may cause cracking. Oil quenching can produce results equal to case-hardening in cyanide baths. It does, however, appear to influence impact strength and residual austenite may have some effect also.

Another Look at Nitriding. H. E. Boyer. (*Steel*, 1956, **139,** Aug. 20, 134–140). The latest techniques in ammonia gas nitriding are noted, and testing methods and recently discovered properties of nitrided steel parts are described.

Five Years' Experience of Sulphinuzation. de Villemeur. (*6th Internat. Mech. Eng. Congress*, 1956, preprint, A2/10, pp. 14).

The Sulf-Inuz Process and the Mechanical Engineer. Caubert. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/3, pp. 12). A process ("sulphinuzation") of obtaining a surface layer rich in S, N and C is described and its peculiar properties discussed. Limiting pressures for failure are estimated.

Heat Treating Retaining Rings for Jet Engines. A. G. Portz. (*Steel Processing*, 1956, **42,** Aug., 467–469). The severity of conditions sustained by compressor blade rings is emphasized. The basic steps in the production of a typical ring are described, together with details of its heat treat cycle.—A. H. M.

Annealing Large Welded Structures of Stainless Steel. H. C. Knerr. (*Machinery*, 1956, **88,** June 8, 961–964). A general discussion on heat-treatment, quenching, handling and bracing methods is given.

A Consideration of the Changes and Mechanisms Involved during the Annealing of Cold-Worked Metals. M. Deighton. (*Bull. Inst. Metals.*, 1956, **3,** July, 93–97). The various changes occurring on annealing, i.e. recovery, polygonization, recrystallization *in situ*, and primary and secondary recrystallization are defined and discussed.—B. G. B.

Unit Stress Relief of Large Drums Aided by Insulation. (*Indust. Heating*, 1956, **23,** Sept., 1930–1936). Practice at the Chicago Bridge and Iron Co. for "in-place" stress relief of large fabricated components is described.

New Extensions to Whitehead's Bright Annealing Plant. (*Metal Treatment*, 1956, **23,** Sept., 356). The installation for bright annealing coiled steel strip totals 37 vertical cylindrical furnaces and 248 annealing containers, being the largest of its kind in Europe. The layout of the plant is described and some features discussed.—A. H. M.

Alloy Wire Fabrication Requires Wide Variety of Annealing Equipment. R. W. Warner. (*Indust. Heating*, 1956, **23,** Sept., 1796–1808, 1832). Nickel alloy and stainless steel wire, rod and strip annealing furnaces are described.

Research on Delay-quenching of the Induction Surface-Hardening. Y. Mitani. (*Tetsu to Hagane*, 1956, **42,** Apr., 313–318). [In Japanese]. If the surface of a steel specimen is austenitised by induction hardening, and quenching by water jet is delayed, the hardness of the specimen is not reduced so much as by immediate quenching: the delay period is several seconds for medium-carbon steel, and scores of seconds for low-alloy steels. Delay-quenching provides a means of stress relief for surface hardness, and is particularly applicable to continuous hardening of long rods.—K. E. J.

The Quenching Properties of Hardening Oils and their Place Among Useful Quenching Agents. A. Rühenbeck. (*Härterei-Tech. Mitt.*, 1956, **10,** 9–27). The author describes useful methods for the testing of the quenching action of cooling media and the various stages of quenching. He goes on to show that maximum quenching intensities of oils are obtained at a bath temperature corresponding to a viscosity of 1·3E. This may be due to the influence of viscosity and flash-point. A quantitative table of various quenching media is given.

High Speed Quenching Oils Meet Broad Range of Heat Treating Requirements. S. J. Barber. (*Iron Steel Eng.*, 1956, **33,** Oct., 168, 171). A brief review.

Mass Marquenching. (*Steel Processing*, 1956, **42,** Oct., 585–588). Treatment of truck transmission parts at International Harvester Co. Fort Wayne works is described. The gas-carburizing system is outlined and illustrated.

Temperability of Steels. L. D. Jaffe and E. Gordon. (*Trans. Amer. Soc. Metals*, 1956, **49,** Reprint 14). A method, based on statistical analysis of the hardness of about 5000 samples of

laboratory heats, has been derived for calculating from the composition of a steel, the temperature required to temper it to a desired hardness after quenching to martensite.—E. E. W.

Isothermal Heat Treating. V. H. Erickson. (*Metal Treating*, 1956, **7**, July-Aug., 2–4). The various methods of isothermal treatment are dealt with and compared with conventional methods. TTT diagrams are used to clarify such procedures as: Isothermal Annealing, Austempering, Martempering and Modified Martempering.—A. H. M.

Austempering. (*Edgar Allen News*, 1956, **35**, July, 137–138). The process and practice of austempering are explained.

The Martempering Process. (*Edgar Allen News*, 1956, **35**, Aug., 165). The process of martempering is explained.

FORGING, STAMPING, DRAWING, AND PRESSING

Recent Trends in Drop-forging Plant. J. H. Callaghan. (*Iron Coal Trades Rev.*, 1956, **173**, Oct. 26, 1017–1020). Details are given of the wide range of plant now available to the drop-forging industry, including board hammers, double-acting hammers, counterblow hammers, high-speed mechanical presses, and forging rolls. Recent trends in design are outlined.—G. F.

The Production of Large Ingots for Forging. E. Bolsi. (*Met. Ital.*, 1956, **48**, Aug. 273–278). [In Italian]. This paper discusses the relative merits of acid O.H. and electric arc steelmaking for large forging ingots. Casting techniques are discussed.—M. D. J. B.

Hot Forging Studies for the Determination of Stability to Change of Form of Die-Forging Steels. F. Wever and W. Lueg. (*Forsch. Wirts. Nordrhein-Westfalen*, 1956, No. 283, pp. 31). Drop-hammer studies on test pieces are described. Temperature dependence is determined for 7 steels and relation to austenite grain size.

The Inspection of Drop Forgings. H. J. Merchant. (*Inspection Eng.*, 1956, **20**, Nov.-Dec., 129–133). Methods used by the Australian Forge and Engineering Pty. Ltd. are described.

Precision Forgings Reduce Machining Costs. L. M. Christensen. (*S.A.E.J.*, 1956, Aug., 34–38). Production of aircraft parts is described and costs are considered.

Automatic Feeding Device for Horizontal Forging Machines. (*Metal Treatment*, 1956, **23**, Sept., 377). The construction and operation of this device are described. Gripping action mechanical movements are operated by air cylinders and the electric controls are synchronized with the action of the forging machine.—A. H. M.

Experiences in Determining Internal Stresses in Large Forgings of Steam Turbine Rotors. O. Keřkovský and O. Žebrák. (*Hutnické Listy*, 1956, **9**, (8), 472–477). [In Czech]. The use of rings sliced from the forging for purposes of estimating internal strains is discussed. The method is recommended only for development work on account of the high cost involved. In correctly heat-treated forgings internal stresses should not exceed permitted levels.—P. F.

New Technique: Lead in Steel Forgings. (*Steel*, 1956, **139**, Aug. 13, 100–101). The use of leaded steel in large gears and forgings is described. The steels, "Hi-Qua-Led," made by Alco Products Inc. by a new process, are available in forged or rolled rings with outside diameters 18 to 145 inches. Improved machinability and other advantages are reported.

Forge Press Maintenance. R. E. Sanford. (*Steel Processing*, 1956, **42**, Aug., 464–466, 474). The author gives detailed information on the action maintenance and repair of brakes, clutches, counter-balance cylinders, friction slip and eccentric shafts. Lubrication, clearances and fastening devices are also discussed.—A. H. M.

Stainless Steel Liner for 7000-ton Forging Press. (*Machinery*, 1956, **89**, Aug. 24, 480).

Tooling Problems of Large Hydraulic Presses. A. E. Faure. (*Steel Processing*, 1956, **42**, Aug., 439–443). The history of the Heavy Forging Press Programme is briefly outlined and some of the reasons for the trend of aircraft manufacturers towards the use of the 35,000 and 50,000 ton presses are given. Some of the problems in the production of large precision forgings are discussed and the aspects of design and construction of dies are considered in more detail.—A. H. M.

Failure of a Shaft due to Blank Forging. E. Hugony. (*Ing. Mecc.*, 1956, **5**, July, 15–16). [In Italian]. The author describes the failure of 2 shafts due to forging without normalization. The chemical compositions of the steels and their structural analyses are given. The author stresses the false economy of the practice.—M. D. J. B.

Heating Methods For Use in Hot Pressing of Screws. S. Hájek. (*Hutník*, 1956, **6**, (7), 205–209). [In Czech]. Heating methods and their relative costs are considered. Electric resistance or induction heating are most appropriate, particularly in mechanized mass production processes.—P. F.

A Study of Lubrication in Relation to Continuous Metal-deformation Processes. L. H. Butler. (*Sheet Metal Ind.*, 1956, **33**, Aug., 571–577; Sept., 647–654; Oct., 727–734, 736). Theories of friction are surveyed, with some detail paid to Bowden welding theory, especially from the metal transfer aspect. Types of lubrication and the special working conditions and lubricant requirements during metal-forming processes are described. Modifications to friction theory and lubrication principles necessary to allow for the bulk plastic flow occurring during continuous deformation are proposed. Some forming processes are considered showing how variation of lubrication conditions may modify forming loads, surface finish, metal flow and other process variables. During continuous deformation it is shown that the properties of the lubricant and the thickness of established lubricant films greatly influence surface finish and deformation may take place via the hydrostatic pressure transmitted by trapped areas of lubricant, resulting in microscopic surface defects which combine to give an overall matt appearance. The work attempts to relate existing knowledge of friction and lubrication to the specific case of metal flow, as opposed to the conventionally considered localized flow at contact points, and quotes the results of some experimental work carried out.—A. H. M.

Closed-circuit Television in Metallurgical Industries. L. Walters. (*Metal Treatment*, 1956, **23**, Sept., 375–376). Several examples of the use of television are described. In the 35,000-ton American press a camera at the press foundation reveals eccentricity of loading. A camera together with BISRA periscope are used to view furnace interior. Colour television and works' intercommunication are briefly discussed.—A. H. M.

Drawn Parts: Save Through Simplified Design. F. Strasser. (*Iron Age*, 1956, **178**, Aug. 2, 88–91). Practical advice on the design of drawn shells and boxes is given, and formulæ specify the flat-blank size for various shapes.—D. L. C. P.

Radial Draw Forming Bends Tough Alloys Easily. E. J. Egan. (*Iron Age*, 1956, **178**, Aug. 9, 75–78). The technique of radial draw forming and its fields of application are described. Stressed sheet, extrusions or sectioned material can be shaped round a forming die by a compression-wipe-shoe to give accurate one-piece parts.—D. L. C. P.

The Manufacture of Thin-wall Tubes. (*Machinery*, 1956, **89**, Oct., 12, 839–843). Methods used by Accles and Pollock Ltd. are described including testing methods.

ROLLING-MILL PRACTICE

Demag Testing Team in Rolling Mill Service. K.-H. Spiller. (*Demag News*, 1956, No. 142, 13–18). A mobile testing unit and its work is described with an account of rolling mill instrumentation and the case of a cold strip rolling reversing mill stand.

The Phenomena Occurring in the Rolling Gap and Their Effect on Roll Force and Torque in Hot Rolling. W. Lueg and H. G. Müller. (*Stahl u. Eisen*, 1956, **76**, Oct.18, 1343–1355). Roll force and torque were measured during production rolling of a number of mild and low- and high-alloy steels on a three-high billet and sheet bar mill. The results obtained are used for a mathematical analysis of the various phenomena involved and presented in tables and graphs. They explain only qualitatively, however, the various forces acting in the rolling gap.—T. G.

Quality and Stability of Rolling-Mill Rolls at the Kuznetsk Metallurgical Combine. I. G. Demko. (*Stal'*, 1956, (5), 459–462). [In Russian]. Data on hardness-distribution over the surfaces of various rolls are given together with indications of roll life. Results obtained cast doubt on the suitability of magnesium-treated cast iron rolls for the finishing rolling of rails and for certain section-rolling operations.—S. K.

Rational Roll-Pass Design for Rails. I. I. Kuchko, M. G. Serkin and L. N. Soroko. (*Stal'*, 1956, (5), 438–445). [In Russian]. Some recently expressed views on the rolling of rails are criticized. It is suggested that the more important parts of the rail should consist of metal from parts of the ingot relatively free from surface defects, and rolling conditions for effecting this are considered.—S. K.

Alloyed Cast Iron Rolls. R. Kamenský. (*Hutník*, 1956, **6**, (7), 201–205). [In Czech]. A review of the recent literature relating to the manufacture, properties and specific applications of all types of cast iron rolls is given.—P. F.

The Segregation of the Special Cast Steel Roll. I. II. T. Kitashima and A. Yoshida. (*Tetsu to Hagane*, 1956, **42**, May, 403–410; 410–415). [In Japanese]. Longer life has been shown by rolls having the composition 0·8% C, 1·0% Cr, 0·3% Mo, etc., and hardened to a Shore number of 40 by spheroidizing, normalizing and tempering. The main factors contributing to cracks were thought to be the techniques of rolling, and segregations during casting. Longitudinal and transverse sections of rolls cast under different thermal and other conditions were examined for segregations, and evidence was collected concerning the solidifying and cooling process.

Heavy-Duty Tapered-Roller Bearings. (*Engineering*, 1956, **182**, Nov. 30, 684–685). Production by British Timkin Ltd. is outlined.

Modern Trends in the Design and Layout of Blooming Mills. D. N. Chatterjee. (*Tisco*, 1956, **3**, Oct., 211–227). The types of layout of modern blooming and slabbing mills are described. The design of soaking pits and equipment used are dealt with. Ingot bogies and turning devices, tables, manipulators, mill housings, screw-down gear, shearing equipment and auxiliary drives are discussed and illustrated by a number of diagrams. The method of calculating torque for main-drive motors is explained and values of the coefficient of friction for steel and chilled iron at various temperatures are tabulated.

Faults in Rolled Billets Caused by Incorrect Heating of Ingots. Z. Volf. (*Hutník*, 1956, **6**, (8), 231–238). [In Czech]. The literature on the subject is reviewed, and the author's own researches are described. Details of the causes and methods of prevention of tear formation are considered in detail.—P. F.

The Industrial Application of Television. (*Metallurgia*, 1956, **54**, Nov., 213–216). A number of examples of the use of closed circuit television are reported. The successful use of television at the Steel Co. of Wales during the failure of a slab manipulator on the slabbing mill is described.—B. G. B.

More Wide Flange Beams. (*Steel*, 1956, **139**, Aug. 13, 106–112). The new plant of the Inland Steel Co., producing wide flange beams (8 to 24 inches) is described and illustrated.

Rationalization of the Design of Machines for Straightening Sections. V. I. Korotkikh. (*Stal'*, 1956, (5), 467–469). [In Russian]. It has been shown that considerable improvements in the design of equipment for straightening rolled sections can be effected, leading to higher productivity without loss of quality.—S. K.

Contour Rolling of Temperature-resistant Aircraft Components. F. J. Altmann. (*Machinery*, 1956, **89**, Sept., 14, 625–631). The mill and shapes produced at the A. O. Smith Corp. Milwaukee are described. Propeller blades are described in particular.

New Sendzimir Mill Rolls Extra-Wide Sheet. (*Iron Age*, 1956, **178**, Aug. 2, 92–93). The working of a recently installed 52 inch Sendzimir mill at Washington Steel Corp., is described. Stainless steel strip up to 48 inches wide can be rolled to thickness tolerances of ± 3%.—D. L. C. P.

The Cooling of Hot Steel Strip with Water Jets. A. Sigalla. (*J. Iron Steel Inst.*, 1957, **186**, May, 90–93). [This issue].

Mechanical Properties of Sheet from Continuously Cast Ingots: Results Obtained at the Krasnoye Sormovo Plant in Russia. (*Iron Coal Trades Rev.*, 1956, **173**, Oct. 19, 973–974). The A.A. Zhdanov two-strand continuous-casting plant in Russia produces 16½ in. × 7 in. slab ingots, and a comparison has been made of the properties of plates rolled from the continuously cast ingots and from conventional ingots. The former are found to have more uniform mechanical properties and, for the same tensile properties, somewhat higher ductility and impact strength values. Internal hot cracks in the cast slabs are welded up during rolling.—G. F.

Planet-Type Rolling Mill Stands for the Hot Rolling of Steel Strip. F. Wiesner. (*Hutník*, 1956, **6**, (8), 225–228). [In Czech]. The construction and performance of planet-type mills described in the literature in recent years are reviewed.

Continuous Strip Gets Tight Coat. (*Steel*, 1956, **139**, Aug. 6, 118–121). Plant for the Armco-Sendzimir hot dip galvanizing process at Jones and Laughlin Steel Corp. is described. Strip is continuously annealed coated and cooled; capacity is 7000–8000 tons per month.—D. L. C. P.

Longitudinal and Cross Rolling of Strip and their Effects on the Quality of Tinning. J. Teindl. (*Sbornik vědeckych praci vysoke školy bańske v Ostrave*, 1956, (1), reprint). [German and Russian summaries]. Sheets 535 × 765 mm were rolled both ways. The cross-rolled sheets had rough, uneven surfaces and took up more tin. Polishing before tinning was tried but was not found advisable. Composition influences amount of tin, in particular steel with 0·12% Si coats thinly, it can also be polished more easily.

Preventing Defects due to Stretching of the Back End of Strip. M. L. Mirenskiĭ. (*Stal'*, 1956, (5), 470–471). [In Russian]. It is sometimes found that the back end of strip during rolling has a smaller transverse horizontal dimension than the rest over a length of 300–500 mm and this can lead to rejects. The causes of this effect and possible remedies are discussed.

Improvements in Hot-Rolled Grain-Oriented Transformer Sheet. A. Mühlinghaus. (*Eng. Digest*, 1956, **17**, Dec., 506–508: from *Electrotechnishe Zeitsch.*, A, **77**, Oct. 11, 732–736). Cold-rolled and hot-rolled sheets are compared from data on specimens and on transformer core trials.

Continuous Tube Rolling. F. Wiesner. (*Hutník*, 1956, **6**, (7), 198–200). [In Czech]. The design and performance of modern continuous tube rolling mills are discussed.—P. F.

The Use of the Modern Assel Mill in Production of Seamless Tubing. C. E. Snee. (*Iron Steel Eng.*, 1956, **33**, Oct., 124–134). The history, principle and operation of the mill is described.

Accurate Method for Measuring the Number of Revolutions on Reducing-Mill Stands. A. K. Shevchuk. (*Stal'*, 1956, (5), 472). [In Russian]. Equipment is described which enables speeds of revolution to be determined with an accuracy of 0·5 r.p.m. It is applied on tube-reducing mill stands and is suitable for other similar operations.—S. K.

Steel Castings for Rolling Mill Couplings. (*Edgar Allen News*, 1956, **35**, June 117–119). The construction and advantages of the "Holset" flexible coupling are described, and the method of determining required coupling size is explained.

MACHINERY FOR IRON AND STEELWORKS

The Special Steelmaker and Power Generation. C. Sykes. (*Cleveland Scientific and Technical Institution, Harold Wright Lecture*, 1956, pp. 35). A review of steam turbine plant with properties of motor shafts and alternator forgings and a section on metallurgical problems with properties of creep-resisting alloy steels is presented.

Guidance on Standard Dimensions of Three-Phase Motors for Intermittent Operation in Iron and Steel Works. H. Schönig. (*Stahl u. Eisen*, 1956, **76**, Oct. 4, 1290–1292). The author discusses the standardization of motors for intermittent operation as in cranes. The new German series of motors have standard overall dimensions for each type and cover the range 4–200 kW at 1000, 750, and 600 r.p.m.—T. G.

Experience with, and Improvements in, Three-Phase Slip-Ring Motors for Intermittent Operation. W. Falck. (*Stahl u. Eisen*, 1956, **76**, Oct. 4, 1292–1295). The author discusses design, construction and electrical details of the new series of slip-ring motors for iron and steel works. Cast iron housings give better cooling than welded steel.—T. G.

Three-Phase Motors with Welded Housing for Intermittent Operation. W. Enders. (*Stahl u. Eisen*, 1956, **76**, Oct. 4, 1295–1297). The author discusses design, construction and electrical details of manufacture of three-phase motors. Switches and resistors are also described.—T. G.

From the First to the Largest Crane Factory. G. Niemann and F. Toussaint. (*Demag News*, 1956, No. 142, 1–12). An illustrated account of cranes of all types including steelworks travelling cranes, and of the firm "Mechanische Werkstätte Harkort and Co.," the parent concern of Demag.

Static Stresses in Electric Overhead Travelling Crane Girders. E. Lightfoot and N. Jackson. (*Structural Eng.*, 1957, **35**, Jan., 1–13). Tests on 7 cranes were made and methods of calculation are given.

LUBRICATION

Solid Lubricants. E. R. Braithwaite. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/13, pp. 11). Surface chemistry is discussed and the properties of graphite and MoS_2, also of plastics and glass. Fretting corrosion is mentioned.

Better Surfaces and Better Sliding Conditions by Use of Molybdenum Disulphide. G. Scherfke. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/12, pp. 6).

Reduction of Coefficient of Friction by the Use of Molybdenum disulphide. P. Bessiere. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/2, pp. 17). High speed and high pressure studies are included, use at high temperatures, surface treatments and use as fretting inhibitor are also discussed.

Applications of Molybdenum Disulphide in the Iron and Steel Industry: Recent Developments and Future Trends. H. P. Jost. (*Iron Coal Trades Rev.*, 1956, **173,** Nov. 2, 1063–1075). The author discusses the fundamental considerations and basic principles of the use of MoS_2 as a lubricant, and gives details of a number of its practical applications in industry. It is emphasised that the compound should be at least 99·3% pure, with the correct particle rise for the particular application, and must be applied correctly.—G. F.

WELDING AND FLAME CUTTING

Progress in the Metallurgy of Welding of Iron and Steel. K. L. Zeyen. (*Schweisstechn.*, 1956, **10,** Aug., 104–108). A review of various test methods is given. The mechanism of brittle fracture is briefly discussed.—U. E.

Inert-Gas-Shielded Metal-Arc Spot Process. J. W. Kehoe and H. J. Bichsel. (*Welding J.*, 1956, **35,** Sept., 895–903). The standard inert-gas-shielded metal-arc welding equipment has been modified so that a controlled time and localized arc spot is obtained. Applications to welding of mild steel, stainless steel and aluminium alloy are given.

Recent Fusion Welding Developments in Soviet Russia. (*Welding Metal Fab.*, 1956, **24,** Nov., 399–403). A description is given of fusion welding processes and modern types of equipment, in particular, automatic machines, used in Soviet Russia.—U. E.

Dec. 1956, p. 477, last abstract, Arc Welding of Aluminium to Steel. Line 2 should read: " by spray-coating the cleaned steel with aluminium and using Al–Si electrodes. A weld of (aluminium sheet 9·52 mm thick to a mild steel bolt 9·52 mm dia. is illustrated)."

Dec. 1956, p. 478, 11th abstract, col. 1. Line 2 should read: " for the bonding of metals with organic adhesives. A study has been made of the effects of the (mechanical properties of the metal)."

In each case a line was dropped.

The Welding of Flake Graphite Cast Irons, with Particular Reference to the Mechanical Properties of Butt-Welded Joints. F. Dunn and G. E. Morton. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1956, **6,** Oct., 364–402). A comprehensive investigation on the mechanical properties of butt welds in six flake graphite cast irons is described, in which the mechanical properties of welds made by the commoner methods for welding cast iron have been compared with those of the same material unwelded. The results of the investigation show that with the exception of the metal-arc welds using a non-ferrous electrode, the metal composition of the parent metal has no significant effect upon the mechanical properties of the weld.—B. G. B.

Resistance Welding and Productivity. J. E. Roberts. (*Welding Metal Fab.*, 1956, **24,** Nov., 404–408). Resistance welding assemblies and design are discussed. The operation and maintenance of these automatic welding machines is shown to be of importance for obtaining satisfactory welded joints. Consideration must also be given to the maintenance of welding electrodes and clamping dies.—U. E.

Submerged Automatic Arc Welding of Rusty Steel. A. G. Mazel and B. P. Vasil'kov. (*Automaticheskaya Svarka*, 1954, **7,** (6), 44–51). Prevention of porosity in welds on rusty metal is discussed. Current, voltage, flux particle size, type of weld, moisture, and C, Mn and Si are all considered and the effects on mechanical properties investigated. Impact values after deformation are given. Rust produces decarburization and increases non-metallic inclusions.

Fusarc/CO_2. A New Welding Process. E. J. Mitchell and W. E. Freeth. (*Welding Metal Fab.*, 1956, **24,** Dec., 431–438). Bare wire gas-shielded methods are reviewed and the new process is described with tables of carbon pick-up and tensile strengths. The new rutile flux is discussed and procedures and the effects of variables are considered.

The Welding of Alloy Steel Tubes for High Temperature and High Pressure Service. (*Sumitomo Metals*, 1956, **8,** Apr., 91–108). [In Japanese]. Recommended welding dimensions and procedures are given for various steel compositions, sizes of tube, wall thicknesses and conditions of service.

The Arc Welding of 2·25% Cr–1·0 Mo Alloy Steel Pipe. R. Huseby and M. A. Scheil. (*Welding J.*, 1956, **35,** Oct., 505s–507s). Discussion of a paper by J. Bland published in April, 1956, **35,** 181s–194s. It is shown that after a year of testing a difference of chromium level existed between deposit and plate and a strong tendency for the carbon to diffuse to the higher chromium alloy. This leads to stress rupture failure.

On the Gas Weldability of Low-carbon Rimmed Steel Hoop; the Results of an Experiment on the Automatic Oxy-acetylene Weldability of ¾-inch Conduit Pipe. Y. Shimokawa, J. Nakagawa and S. Hasebe. (*Sumitomo Metals*, 1956, **8,** Apr., 65–74). [In Japanese]. Results required for assessing the suitability of steels (chemical and mechanical properties given) from three works for automatic production of pipe by oxy-acetylene welding show the effects of impurity content, position of welded seams, hardness, segregations and the actual welding process on weldability.—K. E. J.

Status of Efforts to Develop Non-Manual Pipeline Welding Methods. P. S. Reed. (*Welding J.*, 1956, **35,** Sept., 877–883). The CO_2 welding process is shown to be most suitable for welding pipelines, provided the quality of welding at the bottom of the pipe can be improved to meet requirements. The unit may be an alternative process for small diameter pipe and in emergencies.—U. E.

Welded Reinforcement of Openings in Standard-Steel Tension Members. D. D. Vasarhelyi and R. A. Hechtman. (*Welding J.*, 1956, **35,** Sept., 421s–430s). An investigation is reported on geometric factors such as shape of opening, type and amount of reinforcement and width and thickness of the body plate. Elastic and plastic stress distribution was determined.—U. E.

Problems involved in Spot Welding Titanium to Other metals. F. W. McBee, jun., J. Henson and L. R. Benson. (*Welding J.*, 1956, **35,** Oct., 481s–487s). Properties and problems of spot welding commercially pure titanium to stainless steel and several other metals are examined. It is found that increasing heat caused cracking of the nugget material and formation of an intercrystalline compound in titanium-stainless steel welds. The injection of additional chromium into the titanium-stainless steel weld is found to have little effect on weld characteristics.—U. E.

Arc Welding Costs. A Comparison of Three Firm's Productivity in Making Welds. A. G. Thompson. (*Brit. Welding J.*, 1956, **3,** Nov., 524–542). Time studies were made in three firms using the composite graph as a starting point. A number of indices are developed and used in comparing various aspects of the productivity and two are finally recommended. These are the feet of standard weld (¼ in. fillet) produced (*a*) per net effective hour and (*b*) per clock hour. It has been possible to show that each firm could increase production by upwards of 70% by adopting the best practices, in each respect, of the other two.—U. E.

Hard Facing, Cladded Assemblies, and Plating Obtained by Welding. Gerbeaux. (*6th Internat. Mech. Eng. Congress*, 1956, reprint C1/2, pp. 14). The protection of structural steel and the use of protective linings are discussed, and choice of process is outlined.

Thermal Cutting of Aluminium Stainless Steels, and High Nickel Alloys. E. H. Roper. (*Welding J.*, 1956, **35,** Sept., 915–919). The gas-shielded metal-arc process has been modified for use as a cutting process. It produces cut surfaces suitable in the as-cut condition for further processing, such as welding. The method can be applied to stainless steel, monel, copper, and similar alloys, also aluminium.—U. E.

Transverse Cracking Resulting from Oxygen Cutting. N. N. Breyer. (*Welding J.*, 1956, **35,** Sept., 865–876). Application of general preheating is reported to reduce the occurrence of transverse cracks on the casting side of a cast-armour, rolled-armour weld joint.—U. E.

Methods for Flame Dressing the Metal. M. S. Shlionskiĭ. (*Stahl'*, 1956, (5), 446–449). [In Russian]. Flame-dressing practice at the Kuznetsk Metallurgical Combine is described. Improved results and increased productivity have been obtained by organisational changes.—S. K.

MACHINING AND MACHINABILITY

Metallurgy of Materials for Machining. K. G. Lewis. (*Iron Coal Trades Rev.*, 1956, **173,** Aug. 3, 259–266; Aug. 10, 343–350; Aug. 17, 389–394; Aug. 24, 465–471; Aug. 31, 538–545; Sept. 7, 607–614; Sept. 14, 685–691). The author first discusses the mode of formation of both continuous and discontinuous metal chips during maching. The nature of the cutting process is dealt with, and attention is given to operational factors including shear and clearance angle, heat development and surface temperature of the workpiece, surface finish, and use of cutting fluids. The effect of mechanical properties and steel structure on type of chip formation is described, and data are given on hot machining and high speed machining. The machining characteristics of various steels made by different methods are dealt with briefly, and the effects of structural constituents, grain size, and non-metallic inclusions on machining performance are reviewed. Other factors influencing this performance are considered, including additions to produce free-machining steels and the results of cold working, thermal treatment, and alloying additions. Finally, the special cases of stainless steels, tool steels, cast irons, and non-ferrous materials are considered, and the problems of grinding are briefly surveyed. (208 references).—G. F.

Comparison Between Cutting Speed and Resistance to Wear. Paudrat. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/9, pp. 20). A theory is advanced relating cutting speed and wear resistance and sulphide inclusions are advanced as an example.

The Action of Lubricating Coolants in the Machining of Metals. A. Recine. (*Ing. Mecc.*, 1956, **5,** May, 35–40). [In Italian]. The author discusses the two fundamental purposes of lubricants for cutting, namely to reduce "friction" to a minimum and to provide a good coolant. The results of tests carried out on various metals are described and a number of micrographs showing the effects of lubricating coolants are included.—M. D. J. B.

On the Formation of Chips. G. Carro Cao. (*Ing. Mecc.*, 1956, **5,** Aug., 40–44). [In Italian]. Starting from a theoretical study of chip formation and by applying the theory of plasticity for a wedge penetrating into a semi-infinite plastic mass, the author examines the displacement of the different zones of the chip. The theoretical results arrived at can be experimentally verified by micrographic and micro-hardness tests which bring out the work-hardening of the deformed zones of the cutting.—M. D. J. B.

Break Production Records With Leaded Steels. T. M. Rohan. (*Iron Age*, 1956, **178,** Aug. 2, 83–85). An instance of the machining of leaded steel to make quality pipe couplings, and the advantages and savings from this type of steel, are described.

Machining Studies by Radiometric Methods. F. J. Daasch, S. L. Eisler and W. D. McHenry. (*PB* 111473, 1954; *TIDU List* 617). A Mo- and a B-steel were machined with pile-activated bits and the chip activities measured. Precautions, collecting and sampling methods, and results are given.

Metal Shearing by Means of Abrasive Discs. A. Riggio. (*Ing. Mecc.*, 1956, **5,** July, 41–47). [In Italian]. The author reviews the principles of metal shearing by means of an abrasive disc and describes the results of research carried out in this field. The number of cuts which can be made per hour as well as the size of cut and costs are given.—M. D. J. B.

Tool Consideration in Forming the Austenitic Stainless Steels. L. F. Spencer. (*Steel Processing*, 1956, **42,** Aug., 453–451, 474). The characteristics of austenitic stainless steel and the effect on mechanical properties of cold rolling are outlined. A system of classification of tool materials is proposed and the importance of surface finish in some products is emphasized. The use of materials other than tool steel is discussed; these include carbides, irons, cold rolled steels, cast steels, zinc base alloys, Wood's metal, plastics, wood, etc. Tool steels are grouped as follows: cold heading, forming, wear resisting and high speed steels. Chemical composition, heat treatment and characteristics of these steels are discussed.

Sintox Ceramic Cutting Tool Tips. (*Machinery*, 1956, **89,** Sept. 14, 632–633).

A Study of Drill Bit Wear. E. W. Inett. (*Mine Quarry Eng.*, 1956, **22,** July, 274–280). Rock drills are investigated and recommendations made on design and maintenance.

Electrical Discharge Machining of Form Tools. N. P. Cici. (*Machinery*, 1956, **89,** Sept. 21, 673–677). Spark machining of carbide tips is described.

CLEANING AND PICKLING

Elementary Topics for Research in Metal Finishing. C. James. (*Product Finishing*, 1956, **9,** Apr. 54–62; May 76–89; June, 65–72; Aug., 75–85, 116). In order to stimulate research into electrodeposition and surface treatment, the author systematically discusses each stage and operation of the production of a coating, pointing out problems which remain unanswered. Pretreatments, the electroplating process, the properties of electrodeposited coatings and metal spraying and non-metallic coatings are separately considered.—A. D. H.

Control of Quality in Metal Finishing. A. Heward. (*Electroplating Metal Finishing*, 1956, **9,** Oct., 317–322, 328). The principles of quality control as applied to metal finishing and methods of implementing such systems in metal finishing operations are discussed.—A. D. H.

Deoxidation of Ferrous Metal. The Removal of Scale by Acids and by Sand Blasting. B. Palazzolo. (*Met. Ital.*, 1956, **48,** Aug., 263–269). [In Italian]. The author studies the deoxidation of ferrous metals by means of acids and sand blasting and concludes that for technical and economic reasons sand blasting is preferable. The problem of preventing rust is also examined.—M. D. J. B.

Electrolytic and Chemical Methods of Metal Brightening. H. Silman. (*J. B'ham Met. Soc.*, 1956, **37,** Sept., 417–43). Differences between brightness and smoothness are emphasized and the limitations of both electrolytic and chemical methods outlined. Theoretical consideration of electrolytic polishing are considered and details for polishing stainless steel, chromium, iron and aluminium are given. Various chemical brightening processes are described together with their application to polishing aluminium. Future prospects are briefly discussed.—A. H. M.

Alkaline Derusting Operations. A. C. Hanson. (*PB.* 111744, *TIDU List* 580). The R.I.A. ferrous derusting bath is described, and recommendations for its disposal are given.

Shot Peeners Handle Larger, More Complex Parts. (*Iron Age*, 1956, **178,** Aug. 16, 96–97). Work of a commercial shot peening plant is described. Special equipment has been devised to deal with parts with difficult shapes, or which require selective peening.—D. L. C. P.

Continuous Shot-blasting Equipment for Descaling Steel Sheets. (*Machinery*, 1956, **89,** Nov. 23, 1201–1202). Acid pickling of sheet for Chevrolet cars has been replaced by shot-blasting. Descaling is described.

"High Production" Gun for Vacu-Blast Dustless Shot Blasting Equipment. (*Machinery*, 1956, **89,** Aug. 31, 509–510). Uses are described.

Degreasing of Metal Surfaces. M. Bayon. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C3/1, pp. 8). Methods are outlined, hot and cold, electrolytic and ultrasonic or with combined derusting and passivating.

Flakes in Steel. J. Klička. (*Hutník*, 1956, **6,** (8), 228–230). [In Czech]. The origin and prevention of flakes are discussed with reference to the literature on the subject.—P. F.

Same Setup Tumble—Deburrs Large and Small Parts. J. E. Romais. (*Iron Age*, 1956, **178,** Aug. 2, 86–87).

Roto-finish Processes for Mechanical Grinding, Deburring and Polishing. Boisselet. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/14, pp. 7).

An Unusual Application of Electropolishing. H. E. Z. Gordon. (*6th Internat. Mech. Eng. Congress*, 1956, preprint D/5, pp. 14). Removal of the decarburized layer from heavy-gauge wire is described. A special machine was constructed for the purpose.

Gyrofinishing. G. R. Squibb and F. T. Hall. (*Metal Ind.*, 1956, **89,** Oct. 19, 333–335). An abrasive finishing process, in which the parts are submerged in a rotating mass of abrasive material, is described.

Finishing Certain Parts by Electrolytic Polishing. J. Bergere and R. Mondon. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/1, pp. 13). The process and its effects on fatigue and corrosion are outlined, apparatus and baths described, and electrolytic machining, preparation for plating, and detection of defects dealt with.

Reduction of Wear, and Deburring by Electrolytic Polishing. Heyes. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/8, pp. 11). The use of the acetic anhydride-perchloric acid bath for gear polishing is described. Treatment of drills is also mentioned, and pretreatment for chrome plating.

Automatic Control for Continuous Processing Lines. J. L. Reed. (*Blast Furn. Steel Plant*, 1956, **44,** Dec., 1424). An

account of the control of a pickling, electrolytic cleaning and tinning line with schematic diagrams and circuit details. Winding and shearing are also dealt with.

Ultrasonic Cleaning Units. L. H. Edridge. (*Indust. Finishing*, 1956, **9**, Sept., 101–106). The scope of ultrasonic cleaning methods is assessed and a unit of Swiss manufacture, in which a partial vacuum is applied in the ultrasonic stage is described.—A. D. H.

A Contribution to the Mechanism of Ultrasonic Cleaning. R. Pohlman. (*Metalloberfläche*, 1956, **10**, Sept., 273–276). The mechanism and advantages of ultrasonic cleaning are discussed with the aid of examples.—L. D. H.

The Use of Ultrasonics in the Surface Preparation of Metals. Cleaning with Piezovibrators. P. Wenk. (*Metalloberfläche*, 1956, **10**, Sept., 257–260). The principles of ultrasonic cleaning of metal parts immersed in suitable cleaning fluids are outlined, and examples of its application are given—L. D. H.

Cleaning with Ultrasonic Jet Apparatus. A. Graul. (*Metalloberfläche*, 1956, **10**, Sept., 260–262). The use of jets in ultrasonic cleaning is described. By the use of this apparatus larger amounts of fluid can be used than with quartz vibrators.

Cleaning Plants with Ultrasonics. K. Fürst. (*Metalloberfläche*, 1956, **10**, Sept., 277.) The use of ultrasonics in cleaning plants for various purposes is outlined briefly, and some limitations of the process are indicated.—L. D. H.

Questions of the Use of Ultrasonics in the Surface Treatment of Metals. K. Frölich. (*Metalloberfläche*, 1956, **10**, Sept., 262–265). A summary is given of the contributions to the Colloquium on the Applications of Ultrasonics in the Metal-working Industries, held on 2nd Feb., 1956 at Geislingen/Steige.—L. D. H.

Brittleness of Steel Caused by Hydrogen During Pickling and Electroplating. M. V. Tragardh. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/15, pp. 8). Amsler bend tests were carried out on 120 mm test pieces of different steels, at least 8 tests being made on each. Standardized treatments were applied and results are shown graphically.

Metal Surface Treatment before Painting or Oiling of Metals such as Steel, Aluminium or Zinc. Hess. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C3/6, pp. 11). A brief review.

The Effect of Pickling of Packs on Sheet Surfaces and Tinning. J. Teindl. (*Sbornik vědeckych praci Vycoke školy bañske v Ostravě*, 1955, **1**, (1), 36–48). A bath of sp. gr. 1·22 and containing 10% free H_2SO_4 at 65–75° was satisfactory. Red material blisters less than white. Coating conditions and thicknesses are given.

Skin Protection and Occupational Dermatitis. C. Starkey. (*Electroplating Metal Finishing*, 1956, **9**, Nov., 362–363, 366). Preventive measures for avoiding occupational skin diseases encountered in metal finishing operations are considered in detail.—A. D. H.

PROTECTIVE COATINGS

German Electroplating Industry. A. Kutzelnigg. (*Metal Ind.*, 1956, **89**, Oct. 5, 293–296). A brief review of grinding and polishing, cleaning and degreasing, plating techniques and testing.

Special Applications of Electroplating Processes in a Mechanical Engineering Plant. Organization and Control. Godeaux. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/7, pp. 11). Testing methods and control in collaboration with the laboratory are discussed, and the importance of proper inspection noted.

Report on Some Technological Points Offering a Guarantee of Success in the Design of Surface Treatment Plants. Salvaresi. *6th Internat. Mech. Eng. Congress*, 1956, preprint C3/10, pp. 19). Dip and spray processes are considered and various types of installations evaluated.

Workshop Methods for the Control and Regulation of Electrolytic Baths. Glayman. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/6, pp. 7). A discussion of the data required and the use of the cathometer in control of plating. The cathometer is a spherical electrode of 1 dm^2 area of the metal for plating connected to an accurate ammeter.

The Future of Electroplating Solution Control. K. E. Langford. (*Product Finishing*, 1956, **9**, July, 129–134). The function of the analyst in the plating trade is considered in relation to the need for developing improved methods of bath control. Existing procedures which need improvement are pointed out.—A. D. H.

The Electron Microscope in Electrochemical Research. A. Mohan. (*Product Finishing*, 1956, **9**, Sept., 71–77). The theory and practice of electron microscopy and the application of the instrument to studies of structures of electrodeposits are briefly described.—A. D. H.

Pitting or Porosity of Electroplating Deposits—Their Remedy in Specific Cases. J. M. Odekerken. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/11, pp. 9). Common causes, their detection, and corrective measures are outlined.

New Anticorrosion Plating for Tanks. P. Voigt. (*6th Internat. Mech. Eng. Eng. Congress*, 1956, preprint C1/6, pp. 7). High-alloy thin sheet linings for steel and concrete are described.

Safeguarding the Health of Workers Operating Electroplating Baths or Processes Related to the Preparation. J. M. Odekerken. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/10, pp. 13).

The Effect of Ultrasonics on the Electrolytic Separation of Metals. A. Roll. (*Metalloberfläche*, 1956, **10**, Aug., 230–233). The present state of knowledge concerning the use of ultrasonics in electroplating is reviewed. Particular attention is paid to processes in which the cathode remains stationary, whilst ultrasonic vibrations are directed either normally to or parallel to the cathode. The effect on conditions of deposition and the structure of the single metal or alloy deposit are discussed. (10 references).—L. D. H.

Choice of Electrolytic Coatings for Metals. F. Dyngvold. (*Tek. Ukeblad.*, 1956, **103**, Sept. 13, 747–754). [In Norwegian]. The author discusses the influence of object shape on metal distribution, together with measures to prevent hydrogen embrittlement, loss of fatigue strength and galvanic corrosion. A review is made of metals normally employed for plating, examples being given of their practical use and of those factors which govern the choice of coating in any particular case.—G. G. K.

The Mechanism of Electrodeposition of Bright Nickel. E. Bertorelle' and I. R. Bellobono. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/1, pp. 8). The theory and the effect of additives are discussed.

Chromium Plating. D. J. Fishlock. (*Chem. Indust.*, 1956, Sept. 22, 977–980). A review.

Technical Comments on Some Treatments of Mechanical Components by Hard Chrome Plating. P. Morisset. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A1/10, pp. 9). Uses are outlined with notes.

Thick Deposits as a Means of Integration of Maintenance in Industry. G. D. Piccola and V. Chiaria. (*6th Internat. Mech. Eng. Congress*, 1956, preprint, Section A1/2, pp. 13). Thick chromium and nickel plating methods for machine parts are described and typical parts shown.

The Influence of Chromium-diffusion on the Oxidation Properties of Steels, Nickel and Cobalt-Base Alloys. R. L. Samuel and T. P. Hoar. (*6th Internat. Mech. Eng. Congress*, 1956, preprint B/3, pp. 10). Effects on plain carbon and low and high alloy steels are discussed.

Improvement of Wear Resistance of Ferrous Metal Surfaces by Chromium Diffusion. J. Borgnon. (*6th Internat. Mech. Eng. Congr.*, 1956, preprint A1/1, pp. 12). Methods and results are described. Composition of the coating, essentially carbide, depth and hardness are determined, with a list of uses.

The Development of Nickel and Chromium Plating Since 1945. E. A. Ollard. (*Product Finishing*, 1956, **9**, July, 116–128). Bright nickel baths, the "Kaniger" process for electroless nickel, hard chromium plate and developments in plant and equipment are included in a review of recent changes in nickel and chromium plating.—A. D. H.

Heavy Nickel Electrodeposits. A Remedy for Certain Forms of Fretting Corrosion. Chambaud. (*6th Internat. Mech. Eng. Congress*, 1956, preprint A2/4, pp. 7).

The Mechanical Properties of Nickel and Nickel Alloy Deposits. H. C. Castell. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C2/3, pp. 13). Processes are outlined and the effect of conditions on the properties of the deposit. The effects on the fatigue strength of steel are also considered. Tables of applications and mechanical properties are added.

The Progress of Chemical Nickel Plating in the U.S. and in Europe. Fiedler. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C3/14, pp. 9). An historical review with some account of the baths used.

Electroless Nickel Plate. D. J. Fishlock. (*Product Finishing*, 1956, **9**, Sept., 61–70). Details of the 'Kanigen' process for the deposition of coatings of Ni–8·5% P alloy on steel sur-

faces from alkaline hypophosphite solutions are given. Production methods and applications of the process are described.—A. D. H.

'Kanigen' Chemical Nickel Plating. G. Gutzeit and E. T. Mapp. (*Corrosion Techn.*, 1956, **3,** Oct., 331–336). The authors trace the development of the 'Kanigen' process, in which nickel is deposited by catalytic chemical reduction, and discuss in detail the principles and reactions involved. The composition, structure and properties of 'Kanigen' coatings are dealt with, particular reference being made to the corrosion resistance of such coatings. The authors conclude by indicating the applications of the process. A number of tables is included. (19 references).—L. E. W.

The Deposition of Nickel by Chemical Reduction. A. McL. Aitkens. (*Sheet Metal Ind.*, 1956, **33,** July, 439–443, 455). The "Kanigen" process producing a nickel-phosphorus alloy coating is described. It is claimed that this process overcomes the drawbacks of the Brenner and Riddell method, viz. slow rate of deposition, lack of stability, dull plating and high cost.—A. H. M.

New Non-Electrolytic Plating Process. (*Precision Met. Mold.*, 1956, **14,** June, 66–68). A non-electrolytic process (Kanigen Process) for the deposition of hard non-porous coatings of nickel-phosphorus on iron, copper, aluminium, and their alloys is briefly described.—D. H.

Porosity of Nickel Deposits by Autoradiographic Technique. R. H. Woolff and M. A. Henderson. (*PB* 117368, 1954, pp. 37; *T.I.D.U. List* 531). Steel was plated with radioactive iron and then with nickel from five different baths. A continuity boil test in distilled aerated water is described which correlates well with the radiographic method.

Hot Galvanizing of Small Parts in a Nut and Bolt Factory. A. Gordet. (*Mét. Constr. Mécan.*, 1955, **87,** July, 611–612; Sept., 729–731; Dec., 979–983; 1956, **88,** Jan., 37–39, 51; Mar., 283–285, Apr., 393–395; May, 501–505). A review of the use of zinc coatings on nuts and bolts is presented. A detailed account of techniques used for the cleaning, fluxing, coating and finishing of small parts is presented.—B. G. B.

Protective Zinc Coatings. K. A. van Oeteren-Panhäuser. (*Metall*, 1956, **10,** Aug., 711–715). The protective values of coatings obtained by zinc spraying and zinc loaded paints are surveyed. Details of optimum coating thicknesses and methods of application are quoted, and costs are compared.

The Protection Against Corrosion of Steel by Metallic Coatings, in Particular by Hot Galvanizing. J. F. H. van Eijnsbergen. (*Rev. Soc. Roy. Belge Ing. Ind.*, 1956, Sept.-Oct. 388–393). The relative merits of aluminium and zinc as protective coatings of steel are discussed. The application of zinc by hot dipping, electrolysis, spray, paint, and surface diffusion are considered. The corrosion resistance of zinc coatings are reviewed.—B. G. B.

Heating and Heat Consumption in Galvanizing Furnaces. H. Wübbenhorst. (*Stahl u. Eisen*, 1956, **76,** July 12, 907–913). Possible heating methods, the choice of fuel, the heat balance, development from fuel-fired baths to controlled rotating furnaces, efficiency and heat consumption, and various design types are discussed.

Attack by Iron-Saturated Liquid Zinc on Iron Containing Copper. D. Horstmann. (*Arch. Eisenhüttenwesen*, 1956, **27,** Apr., 231–233). The time- and temperature-dependence of the attack are dealt with for iron containing up to 0·91% copper. The structural composition and growth of the iron–zinc alloy layers is discussed.

Velindre Starts Production. (*Tin*, 1956, Sept., 195–196). An account of the Steel Company of Wales plant.

Aluminizing Progress. H. G. Erstrom. (*Light Met. Age*, 1956, **14,** Aug., 24–27). Hot-dipping by the Mollerizing process is described, with notes on uses and tests.

The Structure of Metal Spraying Layers. A. Matting and W. Raabe. (*Schweissen u. Schneiden*, 1956, **8,** Oct., 369–374). The temperature of sprayed metal particles on the target was determined by using optic-photographic and calorimetric methods. It was found that the temperature of the particles when impinging on the surface is the same as the solidus temperature of the alloy used in the wire. An investigation on the constitution of the particles showed that cohesion exists between metal particles and the metal surface, occasionally with oxides between the two layers.—U. E.

Flame Spraying of Metals and Plastics. G. R. Lawrance. (*Corrosion Techn.*, 1956, **3,** Aug., 254–257). The author first describes briefly the pistol mechanism involved in the Schori process and then discusses the surface preparations that prove adequate for satisfactory metallizing. Typical commercial metal and plastic spraying equipment is described. The applications of zinc, aluminium, polythene, nylon, and 'Thiokol' sprayed articles are dealt with.—L. E. W.

Stabilization of Passive State in Austenitic Chrome-Nickel Steels by Anodic Polarization in Boiling 4n Sulphuric Acid. C. Carius. (*Arch. Eisenhüttenwesen*, 1956, **27,** June, 401–412). Potential measurements were made on a steel containing 0·05% C, 18·1% Cr, 0·4% Mo and 10·1% Ni in sulphuric acid at room and boiling temperatures. Voltage-current, voltage-time and voltage-temperature curves are given and the results are discussed.

Cement Linings for Iron and Steel Pipes. J. Rutle. (*Tek. Ukeblad*, 1956, **103,** June 28, 597–600). [In Norwegian]. Details are given of American codes of practice and results obtained with cement-lined water pipes made of iron and steel so as to prevent leakage due to electrolytic corrosion. Hints are given on lining thickness, cement mixing and treatment of new and old pipes. Tests carried out on lined pipes showed that under severe stresses two layers of cement were necessary to prevent film disruption, but otherwise, the conclusions drawn were favourable.—G. G. K.

Synthetic Materials for Metal Coating. G. Schulz. (*Schweiz. Arch. Weiss. Techn.*, 1956, **22,** June, 178–182). A general description of the plastic materials used for metal coating is given with some detail on the different means of application and the protective qualities of the coatings.—J. R. P.

Practical Aspects of Synthetic Enamelling. F. Westall. (*Engineer Foundryman*, 1956, **20,** March, 41–43, 64). Details of pretreatment processes and methods of applying synthetic enamels are described.—A. D. H.

Organic Coatings Used in Cathodically Protected Domestic Hot Water Tanks. B. G. Brand, G. A. Hudson, F. W. Fink and E. R. Mueller. (*Indust. Finishing*, 1956, **9,** Aug., 51–53). The results of thermal shock, alkali resistance, applied potential and magnesium anode tests applied to epoxy and phenolic resin coatings to assess their value as protective coatings for domestic hot water tanks are presented. It is concluded that the coatings may be suitable for this application.—A. D. H.

Current Mill-Room Practice. S. H. Ryder. (*Institute of Vitreous Enamellers Bulletin*, 1956, **6,** March, 93–100, 102). Wet milling in ball mills, fineness, and viscosity measurements, application weight, and the use of high-density grinding balls are considered with notes on two U.S. plants.

Role of Nickel Dip in Enamelling of Sheet Steel. D. G. Moore, J. W. Pitts and W. N. Harrison. (*Indust. Finishing*, 1956, **9,** Aug., 36–41). The effects of a flash coat of nickel on the development of adherence and fish-scaling tendency of a typical ground coat enamel on enamelling iron and titanium stabilized low-carbon steel were examined. The beneficial effect of the pretreatment was confirmed and the improved adherence is considered to be connected with the roughening of the surface during the nickel dip and during the firing of the enamel.—A. D. H.

The Use of Gas in Vitreous Enamelling. H. Capper. (*Indust. Gas*, 1956, **19,** June, 287–306, 315). The various stages of the vitreous enamelling process including enamel frit production, metal preparation and base coat and cover coat application and treatment are described with specific reference to a typical production plant.—D. H.

The Use of Gas in Vitreous Enamelling. (*Indust. Finishing*, 1956, **9,** July, 676–683). A semi-mechanized enamelling works, Elm Works Ltd., London, is described in detail. Enamel frit production, metal preparation, base coat application and treatment, cover coat application, and treatment and inspection are dealt with.—A. D. H.

Works Practice in Coloured Sheet-Iron Enamelling in Yugoslavia. H. Ankerst. (*Metal Finishing J.*, 1956, **2,** June, 179–184). Experience (*a*) with coloured cover-coat enamels, (*b*) in decorating white titanium enamels and (*c*) in special enamelling procedures obtained in a Yugoslav works are presented.—A. D. H.

Historical Notes on Enamelling Iron for Porcelain Enamel. J. J. Canfield. (*Products Finishing*, 1956, **20,** Aug., 30–32). A brief note on the development of the use of Armco iron for enamelling.—A. D. H.

Suggested Method of Test for Set Characteristics of Porcelain Enamel Clays. (Slump Method). A.S.T.M. Subcommittee IV. (*Amer. Soc. Test. Mat. Proceedings*, 1955, **55,** 429–430).

Detergent Corrosion Test for Vitreous Enamel Surfaces.

J. C. Harris, M. G. Kramer, and M. V. Trexler. (*Amer. Soc. Test. Mat. Bull.* 1956, Sept., 61–64). A test procedure is described for determining the corrosive effects of water soluble cleaners on vitreous enamel surfaces.—B. G. B.

Measurement of Gloss in Vitreous Enamels. (*Metal Finishing J.*, 1956, **2**, 314–316). The development of a goniophotometer for the standardization of gloss-meters is described.—A. D. H.

An Investigation into the Viscosity of Enamels. N. S. C. Millar and C. H. Buck. (*Metal Finishing J.*, **2**, May, 151–156, 160). A rotating cylinder viscometer used to determine the viscosity of enamels at 750°–900° C is described. The effect of the addition of oxides to AR titania opacified enamels on the firing range, flow and fusion point was studied and correlated with the corresponding change in viscosity. The results were used to improve the properties of another titania enamel to which additions of silica were made.—A. D. H.

Some Observations on the Incidence of Fishscales on Enamelled Steel Pots. E. Messmer. (*Metal Finishing J.*, 1956, **2**, Sept., 335–344). A systematic series of observation on fishscales in two types of enamelled pans for use on electric and gas stoves respectively is analysed. It was found that the gauge of the sheet and its quality were the main factors influencing the incidence of the defect.—A. D. H.

Electrochemical Studies of Paint Films on Metals. D. M. Brasher. (*Electroplating Metal Finishing*, 1956, **9**, Sept., 280–284). Changes in the properties of paint films were studied by electrode potential, electrical resistance and capacitance methods which are described. The last method was used to determine the effect of metallic benzoates as inhibitive pigments in sea water.—A. D. H.

Behaviour of Bituminous Lacquers Applied on a Steel Arch during 18 Years of Exposure to the Weather. Blancheteau. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C4/1, pp. 10).

Properties, Specifications, Tests and Recommendations for Coal Tar Coatings. Part 1—Hot Applied Coatings. W. F. Fair, jun. (*Corrosion*, 1956, **12**, Nov., 579–587). The properties required of hot-applied coal-tar enamels suitable for protecting iron and steel pipes both internally and externally are described. A description of the techniques of application is followed by details of suitable specification tests. Enamels should be chosen from material of similar physical properties to types of known previous good performance.—J. F. S.

Use of Baked Coatings in the Oil, Chemical and Processing Industries. G. J. Duesterberg. (*Corrosion*, 1956, **12**, Nov., 576–578). Actual and potential uses of baked phenolic and epoxy-modified phenolic resin coatings to combat corrosion are reviewed.—J. F. S.

An Improved Method of Non-Metallic Anticorrosive Spray-gun Coating. P. Voigt. (*6th Internat. Mech. Eng. Congress*, 1956, reprint C4/8, pp. 11). Equipment and control are described. A high polymer powder is used.

Notes on Plastic Coatings by Oxy-Acetylene or Oxy-Propane Spray-gun Application. Cauchetier. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C4/2, pp. 4).

The Use of Hot-Dip Strippable Coatings for the Protection of Spares. F. A. S. Wood and N. R. Kirkby. (*Corrosion Techn.*, 1956, **3**, Nov., 355–356). The properties, advantages and applications of hot-dip strippable coatings (H.D.S.C.) are outlined. The precautions to be observed to ensure satisfactory coating are indicated.—L. E. W.

Tank Protective Coatings. (*Corrosion Prevention and Control*, 1956, **3**, Dec., 55). The application of Tanclene anti-static interior tank protective paint (produced by the Red Hand Compositions Company) is described and its protective properties illustrated by data from immersion tests. A three-coat treatment is applied and required five days to cure. The curing time can be reduced to six hours by heating with infra-red to provide a surface temperature of 100°–120° F.

Silicone Protective Coatings in Metal Finishing. P. A. J. Gate. (*Product Finishing*, 1956, **9**, Aug., 48–51). The use of paints containing silicones as finishes for steel is discussed. The coatings which require stoving at 200–250° C are protective up to 1000° F.—A. D. H.

The Decade's Developments in Vitreous Enamelling. J. C. Oliver. (*Product Finishing*, 1956, **9**, July, 172–180). New production methods, improved formulations and new applications for vitreous enamel are reviewed.—A. D. H.

Enamel as a Protection Against Corrosion. J. Fichte. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C4/3, pp. 19). A review of corrosion processes, the composition and application of vitreous enamel, and its resistance to liquids and to atmospheric corrosion.

Shot Blasting of Cast Iron for Vitreous Enamelling. T. O. G. Wilkes. (*Inst. Vitreous Enamellers Bull.*, 1956, **7**, Sept., 3–6). The manufacture of chilled shot for cleaning iron castings, and machines used for the process are described.—A. D. H.

Fundamental Thermal Deflection Analysis of Enamel-Metal Systems. J. H. Lauchner, R. L. Cook and A. I. Andrews. (*J. Amer. Ceram. Soc.*, 1956, **39**, Aug., 288–292). The deflection-temperature relationship of enamelled metal strips was automatically recorded on a function plotter from 25° to 600° C. The thermal expansions of enamels and metal were correlated with the observed deflections produced by the composite strip. Metal and coating thicknesses were related to deflection, and corresponding stress values were calculated.

Effect of Temperature on Modulus of Elasticity of Porcelain Enamelled Steel. R. E. Cowan, A. W. Allan, A. L. Friedberg. (*J. Amer. Ceram Soc.*, 1956, **39**, Sept., 293–300). The moduli of elasticity of porcelain enamelled steel specimens were measured at different temperatures by a sonic method Equations and constants are presented by which the elastic constant can be calculated for any enamel thickness.—D. L. C. P.

Rheem Glass Lines its Heaters. (*Steel*, 1956, **139**, Oct. 15, 124–125). Improvements in fabrication and enamelling of water heaters are described.

The Role of Oxygen in Iron-Enamel Adherence. A. G. Eubanks and D. G. Moore. (*Ceramics*, 1956, **8**, July, 118–119). Samples were fired in a sealed furnace to which known N_2 and O_2 concentrations were admitted. Cobalt oxide-containing enamels attain maximum adhesion at specific O_2 pressures.

Chipping. S. E. A. Ryder. (*Inst. Vitreous Enamellers Bull.*, 1956, **6**, May, 108–112). The seriousness of the problem of chipping of vitreous enamel is considered. It is suggested that the application of thinner coatings will help to reduce the tendency to chipping and that the industry should undertake co-operative research into the direct application of cover coat to steel.—A. D. H.

Anodic Polarization of Steel before Painting. J. Frasch. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C4/4, pp. 5). Chromate pretreatment, particularly of underwater structures, is discussed.

15 Years Experience of Hot [Paint] Spraying in Sweden. K. Liedberg. (*6th Internat. Mech. Eng. Congress*, 1956, preprint C1/4, pp. 15).

Pre-Painted Strip. (*Electroplating Metal Finishing*, 1956, **9**, Nov., 367–368). The lay-out of an American plant for the production of paint-coated steel strip on a continuous basis is described.—A. D. H.

Zinc Rich Paint: A Coating for Problem Surfaces. D. E. Wilbur. (*Iron Age*, 1956, **178**, Aug. 9, 82–84). The properties and applications of zinc paints are described. A non-settling paint is available which gives a dry film containing over 95% zinc. Properties compare favourably with galvanized coatings, and rusty and scaled surfaces need only be brushed before application.—D. L. C. P.

Protecting Steelwork from Rust. (*Corrosion Techn.*, 1956, **3**, Dec., 393–396). In this article are presented briefly the results of surface treatment tests carried out by Lewis Berger in co-operation with Jenolite Ltd. Humidity cycle, marine cycle and immersion tests were conducted and it was found that pretreatment with Jenolite RRN greatly increased the resistance to rusting under all three conditions of exposure for the four priming paints used. These were (*a*) the Lewis Berger 'Leadium' metallic lead paint No. A.100, (*b*) the quick-drying version No. A. 859, (*c*) a paint conforming to T.S. 226A, and (*d*) a ships' bottom primer.—L. E. W.

POWDER METALLURGY

Powder Metallurgy. M. Chubachi and T. Fujiyama. (*Sumitomo Metals*, 1956, **8**, July, 185–200). [In Japanese]. A review covers methods of preparation of ferrous and non-ferrous metallic powders, methods of sintering, and the properties of sintered parts, particularly metallic materials and self-lubricating porous bearings. (17 references).—K. E. J.

The Fundamental Properties of Metals and Powder Metallurgy. H. W. Greenwood. (*Metallurgia*, 1956, **54**, Oct., 187–188). The improvement in the properties, particularly ductility, of certain metals (including iron) as small amounts of impurities are removed and the beneficial effects of grain

coatings in the production of powdered 18-8 stainless steel are considered.—B. G. B.

The Process of Sintering. G. Carro Cao. (*Met. Ital.*, 1956, **48,** July, 225–233). [In Italian]. The author reviews the principal studies on sintering and describes in detail experimental studies which throw light on the phenomenon and notably on the effect of grain size and pressure of packing. The studies were carried out by means of X-ray diffraction on different powder samples with varying grain size and submitted to different pressures. A number of micrographs are shown.—M. D. J. B.

Quality Control of Powdered Metal Parts. G. L. Bachner. (*Machinery*, 1956, **89,** Nov. 16, 1128–1130). Special tests used by Yale and Towne Powdered Metal Products Division are reviewed. Size, strength in torsion, crushing strength and ductility testing are mentioned.

The Deformation Characteristics of Sintered Iron. P. Fischer and A. Steinegger. (*Schweiz. Arch. Wiss. Techn.*, 1956, **22,** Sept., 290–293). The effects of compression and elongation in improving the porosity and hardness characteristics of sintered iron are described.—J. R. P.

FERRITES, CERMETS, AND CARBIDES

The Crystal Structure of Calcium Ferrite and β-Calcium Chromite. P. M. Hill, H. S. Peiser and J. R. Rait. (*Acta Crystallographica*, 1956, **9,** Dec. 10, 981–986).

The Structure of Tetragonal Copper Ferrite. E. Prince and R. G. Treuting. (*Acta Crystallographica*, 1956, **9,** Dec. 10, 1025–1028).

Nickel Copper Ferrites for Microwave Applications. L. G. Van Uitert. (*J. Appl. Phys.*, 1956, **27,** July, 723–727). Density-firing temperature and resistivity and magnetic saturation—composition curves are determined.

Magnetic Properties of Some Orthoferrites and Cyanides at Low Temperatures. R. M. Bozorth, H. J. Williams and D. E. Walsh. (*Phys. Rev.*, 1956, **103,** Aug. 1, 572–578). Apparatus is described and measurements made down to 1·3° K.

Formation of Manganese Ferrite by Solid-State Reaction. H. H. Kedesdy and A. Tauber. (*J. Amer. Ceram. Soc.*, 1956, **39,** Dec., 425–431). Optical, X-ray, analytical and thermal analysis methods were used and also magnetic studies. The initial formation of solid solutions was examined in detail.

Method of Forming Large Ferrite Parts for Microwave Applications. L. G. Van Uitert, F. W. Swanekamp and F. R. Monforte. (*J. Appl. Phys.*, 1956, **27,** Nov., 1385–1386). A pressing and sintering technique is described.

Cation Distributions in Ferrospinels. Theoretical. H. B. Callen, S. E. Harrison and C. J. Kriessman. (*Phys. Rev.*, 1956, **103,** Aug. 15, 851–856). **Magnesium–Manganese Ferrites.** C. J. Kriessman and S. E. Harrison (857–860).

An Impact Test for Use with Cermets. E. J. Soxman, J. R. Tinklepaugh and M. T. Curran. (*J. Amer. Ceram. Soc.*, 1956, **39,** Aug., 261–265).

PROPERTIES AND TESTS

Probability and Operational Research. F. Benson. (*Research*, 1956, **9,** Sept., 329–334). Applications in steelworks are mentioned.

Energy Integrals for 3D Electrons in Body-centred Iron. M. Suffczynski. (*Acta Phys. Polonica*, 1955, **14,** (6), 493–495).

On the Mechanism of Intercrystalline Cracking. C. W. Ohen and E. S. Machlin. (*Acta Met.*, 1956, **4,** Nov., 655–656). A letter in which experimental evidence is presented to support the view that grain boundary shear is a necessary condition for the formation of intercrystalline voids.—A. D. H.

Structure and Properties of Metals. R. Fieschi and M. Tosi. (*Ing. Mecc.*, 1956, **5,** July, 49–55; Aug., 57–62; Sept., 5–8). [In Italian]. The authors study the properties of atoms and interatomic forces. The theory of electronic wave bands provides an explanation for the electric and thermal properties of matter and notably the difference between conducting and non-conducting materials. The main differences between ideal and real materials are discussed. Amongst the imperfections found in real solids are dislocations. The part played by these in explaining the mechanical properties of metals is examined. Substitutional and interstitial alloys and in particular steels are studied.—M. D. J. B.

The Behaviour of Metals Under Stress. (*J. Australian Inst. Met.*, 1956, **1,** Oct.). Report of a symposium. **Introduction.** A. R. Edwards. (89). **Lattice Defects and Plastic Deformations.** L. M. Clarebrough. (90–111). Dislocation theory and its relation to strength is outlined. (73 references). **The Plastic Deformation of Single Crystals.** N. S. Paterson. (112–124). Slip and strain hardening and deformation by twinning are described. (115 references). **The Deformation of Polycrystalline Aggregates.** M. E. Hargreaves. (125–133). Attempts at calculation are outlined and a mechanism is discussed. **Slow Deformation of Metals Particularly at Elevated Temperatures.** R. C. Gilkins. (134–147). Creep curves are shown and discussed and theories outlined. Observed structural changes are then described and their relationship to the empirical equations considered. Fine slip and boundary sliding are illustrated and mechanisms and models outlined. More recent theoretical treatments are then mentioned. (76 references). **Fatigue.** A. K. Head. (148–154). Fatigue mechanism and crack initiation in simple metals are discussed. **Fracture.** H. L. Wain. (155–169). Theories of fracture leading to the Mott–Stroh model are discussed. Observations on single crystals and polycrystalline metals are then briefly considered and the effects of impurities. The present position is then summarized. (59 references).

Plastic Deformation of Metals. A. Hentz. (*Jernkontorets Ann.*, 1956, **140,** (8), 575–602). [In Swedish]. A survey is made of the main structural and crystallographic features of plastic deformation, with special reference to conditions prevailing in pure metals at room temperature and normal strain rates in simple stress situations. The author presents his conclusions concerning slip in single crystals and polycrystalline metals, reorientation from slip (lattice bending and rotation, deformation bands, etc.) and twinning.—G. G. K.

Slip, Twinning and Fracture in Single Crystals of Iron. J. J. Cox, G. T. Horne and R. F. Mehl. (*Trans. Amer. Soc. Metals.*, 1956, **49,** Preprint 1). The properties were investigated at temperatures from 392° F to − 321° F. A critical shear stress law was found to be a valid criterion for slip at all temperatures. For twinning, however, the direction of the applied stress was important. Crystals whose nearest (001) pole fell less than 55° from the tensile axis fractured in cleavage and showed a wide variation from the normal stress law.—E. E. W.

Atomic Defects in Crystal Structure. A. Steinemann. (*Schweiz. Arch. Wiss. Techn.*, 1956, **22,** July, 226–237). A review of the primary imperfections in crystal structure of solids with a description of the properties associated with these imperfections.—J. R. P.

A Complex-Variable Approach to Stress Singularities. M. L. Williams. (*J. Appl. Mech.*, 1956, **23,** Sept., 477–478). See Huth, *ibid.* 1953, **75,** 561. Two approaches to an example are shown to be in agreement.

A Method for Calculating Stress-Concentration Factors. M. Hetényi and T. D. Liu. (*J. Appl. Mech.*, 1956, **23,** Sept., 451–457). Filleted and notched bars are dealt with on a few simple assumptions concerning load distribution.

Studies on Plastic Flow of Anisotropic Metals. L. W. Hu. (*J. Appl. Mech.*, 1956, **23,** Sept., 444–450). A mathematical treatment is given for plastic stress-strain relations with strain hardening. Yielding in a thick-walled cylinder is considered and effects on biaxial tension–tension and tension–torsion tests.

Internal Friction Methods applied to Metallurgical Problems. W. R. Thomas and G. M. Leak. (*Metal Treatment*, 1956, **23,** Sept., 359–366; Oct., 413–419). The authors give an introduction to the subject of internal friction in metals and show how the phenomenon can be measured experimentally. A number of examples are discussed in detail to indicate the type of problem which might fruitfully be investigated.—A. H. M.

Theory of Mechanical Damping Due to Dislocations. A. Granato and K. Lücke. (*J. Appl. Phys.*, 1956, **27,** 583–593). A theory provides a quantitative interpretation of the hysteresis loss, which is found to be independent of frequency over a wide range, and which has a strain-amplitude dependence of the type observed in the kilo-cycle range. A frequency-dependent loss is also found which has a maximum in the high megacycle range.—E. E. W.

Researches on Damping in α-Iron Annealed or Cold-worked, in the Presence of Small additions of Carbon and Nitrogen. W. Köster. (*Publ. Assoc. Ing. Faculté Polytechn. Mons*, 1956, (3), 1–13). A torsion pendulum apparatus is described and ageing and carbide and nitride precipitation correlated with its indications at 40° and 200°. Armco iron and basic Bessemer steel are used. Micrographs are shown.

Dynamic Elastic Moduli of Iron, Aluminium and Fused

Quartz. D. S. Hughes and C. Maurette. (*J. Appl. Phys.*, 1956, **27,** Oct., 1184–1186). The dynamic moduli of Armco iron, etc. can be represented by a linear function of pressure and temperature. By using thermodynamics, the static bulk moduli of other observers have been used to compute the equivalent adiabatic moduli and comparisons show that at low pressure the dynamic modulus of Armco is less than the adiabatic.—E. E. W.

Effect of Silicon on Transverse Properties and on Retained Austenite Content of High Strength Steels. J. Vajda, J. J. Hauser and C. Wells. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 24). Transverse tensile and transverse impact properties of both fully-quenched and slack-quenched AISI 4340 steel containing added silicon (1% and 1·5%) were investigated. Si markedly increased the amount of austenite retained, especially in slack-quenched structures. Results are given for the relation between tempering temperature and retained austenite. The effects of multiple tempering and refrigeration on retained austenite are described.—E. E. W.

Some Exploratory Observations of the Tensile Properties of Metals at Very Low Temperatures. E. T. Wessel. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 3). The tensile properties and flow characteristics of several metals, including steels SAE 4340 and 3335, were investigated over a wide range of low temperatures down to 4·2° K. The results are given graphically.—E. E. W.

Effect of Alloying Elements on the High-temperature Tensile Strength of Normalized Low-carbon Steel. J. Glen. (*J. Iron Steel Inst.*, 1957, **186,** May, 21–53). [This issue].

The Deformation and Rupture of Gray Cast Iron. W. R. Clough and M. E. Shank. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 8). Thin walled combined stress specimens, immersion type density measurements and microscopical methods were used for the investigation of mechanical properties. During plastic deformation there is a density decrease which is a smooth function of applied stress and which is greater for biaxial than for uniaxial stressing. The plastic lateral contraction ratio increases with increasing tensile stress but is only about half of the classical value. Fracture occurs in a direction normal to the greatest tensile stress and is mainly through graphite. Other features are noted and explanations advanced.—E. E. W.

The Influence of Temperature on Pre-Yield and Anelastic Microstrain in Low Carbon Steel. J. A. Hendrickson, D. S. Wood and D. S. Clark. (*Acta Met.*, 1956, **4,** Nov., 593–601). The pre-yield plastic and anelastic microstrain in annealed low carbon steel was measured at − 31°, 24° and 66° C over a range of rapidly applied stresses. The rates of pre-yield microstrain were compared with those predicted on the basis of dislocation theory and it is concluded that the existing theory requires modification.—A. D. H.

Effect of Strain Rate on the Tensile Properties of SAE 4340 Steel. R. F. Klinger. (*PB* 111948, 1955, pp. 30; *TIDU List* 578). The range from 0·00002–20·0 in./in./sec was covered. Strength is greater at higher rates and the differences are less as the strength level of the steel increases. This trend continues to 220,000 psi which shows no change at higher rates. Elongation is unchanged.

Stress Analysis by Photo-Elastic Varnishes. F. Zandman. (*Acier Stahl Steel*, 1956, **21,** Sept., 356–364). Theory is given and uses illustrated.

On the Accuracy of Strain Determination by Brittle Coatings. T. Nishihara, S. Taira and H. Maeda. (*Tech. Rep. Eng. Res. Inst. Kyoto Univ.*, 1956, **6,** Feb., 1–12). [In English]. Investigations were made of a method of determining strain by means of a coating of synthetic resin sprayed on to the surface of a metal specimen and dried at 50–60° C for several hours, giving a coating about 0·10 mm thick. On applying strain, the value when the first crack appears is the critical strain, and the surface strain distribution is investigated from the subsequent pattern of cracks. Three cases were examined: (*a*) when the loading speed varied, (*b*) when the strain gradient was in the direction of the tensile strain, and (*c*) when the gradient was at 90° to the direction of tensile strain. The method gives high results in all cases, and methods of correction of the experimental values are described.—K. E. J.

The Strength of Certain Stainless and Carbon Steels at Low Temperatures. S. C. Collins, F. D. Ezekiel, O. W. Sepp and J. W. Rizika. (*Amer. Soc. Test. Mat. Preprint. No.* 74, 1956). Preliminary results of torsion and tension tests of various steels at very low temperatures are reported. The tests are conducted in a stream of cold gaseous helium whose temperature can be adjusted to any value between 10 and 300° K. Unusual discontinuities in the stress-strain relationship of stainless steel at lower temperatures are observed. The effect of work hardening at low temperature upon the room-temperature strength is also noted.—B. G. B.

Notch Ductility of Type 410 (12% Cr) Stainless Steel. F. A. Brandt, H. F. Bishop, and W. S. Pellini. (*Trans. Amer. Soc. Metals*, 1956, **48,** 368–390). The factors influencing the resistance to brittle fracture of Cr steels have been investigated by Charpy **V** tests and by sharp crack tests of large specimens involving drop-weight and explosion loading. The addition of 0·7% Mo and Ni and the lowering of Si to approximately 0·5% gave a stainless steel which was highly resistant to brittle fracture at low winter temperatures.—E. E. W.

Effect of Composition and Heat Treatment on Impact Properties of AISI Types 403 and 410 Stainless Steel. E. J. Whittenberger, E. R. Rosenow, and D. J. Carney. (*Amer. Soc. Test. Mat. Preprint No.* 75, 1956). The use of higher carbon contents (0·12 to 0·19%), the addition of nitride- and carbide-forming elements (aluminium, vanadium, and titanium), and austenitizing at lower temperatures to achieve finer grain size and minimize intergranular carbide precipitation during subsequent tempering, were found to improve the impact properties of the 12% chromium steels investigated. Increasing amounts of manganese, nickel, and nitrogen did not appear to increase the impact strength.—B. G. B.

On the Methods of Deformation in Metals. L. Habraken and T. Greday. (*Rev. Univ. Min.*, 1956, 9th series, **12,** Aug., 209–227). A comprehensive review of the present knowledge of the deformation of single crystal and polycrystalline metals is presented and discussed in relation to the theory of dislocations. The role of imperfections in the crystal lattice on the deformation is reviewed and different forms of deformation considered. (114 references).—B. G. B.

The Stress-Hardness Relation. A. Dervishyan. (*Amer. Soc. Test. Mat. Bull.*, 1956, July, 71–75). A theoretical and experimental study of the stress-hardness relation is presented. The theoretical study is based on the Herz–Huber theory and it is assumed that yielding occurred when the shear-stress reached a maximum. Experimental results show a linear relationship between stress and hardness, as defined. The possibilities of the techniques being used to determine the stress in structures already built is discussed.—B. G. B.

Deformation and Recrystallization of Silicon Iron. G. Wiener and R. Corcoran. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206;** *J. Met.*, 1956, **8,** Aug., Section 2, 901–906). The authors have studied the orientations and microstructures of 3% Si–Fe alloys after deformation and recrystallization, and observe that the components found after deformation agree with recent work on single crystals. Temperature of recrystallization is found to have a pronounced effect on microstructure and orientation, being interpreted in terms of boundary angles. No simple angular relation between the deformation and high temperature recrystallization textures is detected.—G. F.

Contribution to the Calorimetric and Metallographic Study of Work Hardening Energy Caused by Cutting a Workpiece. F. Eugène. (*Microtecnic*, 1956, **10,** (3), 114–119; Discussion 119). Cooling of chips by radiation was found to be too great for accurate work so that only the tool and workpiece were studied and the energy accumulated in the chip calculated by difference. Steels of 0·1, 0·2, 0·4 and 0·55% C were used and energy related to cutting speed and rate of feed. A microhardness study of the work-hardened layer was made and crushing tests and measurement of tangent cutting force were also applied. Work of elastic deformation was similarly computed.

The Effects of Working on the Bond Strength of Crystals of α-Solid Solution of Iron–Nickel Alloys. V. K. Kritskaya, G. V. Kurdyumov, and L. V. Tikhonov. (*Doklady Akad. Nauk, SSSR.*, 1955, **102,** 271–274). A 4% nickel alloy was annealed and examined with X-rays. Forging, deformation and quenching after annealing all changed the bond strength.

Effect of Sulphur in Furnace Atmosphere on Deformation Behaviour of Unalloyed Steels. K. Born. (*Stahl u. Eisen*, 1956, **76,** June 28, 788–799). Hot bending tests were carried out after heat treatment in sulphurous and non-sulphurous furnace gases. Material very low in Cu and Sn and other electro-, Thomas, and Siemens-Martin steels with graded Cu and Sn contents were used. The effects of Cu and Sn

on fracture, and of sulphur-bearing gas on the state of surfaces, are discussed. Evidence is given to show that surface enrichment of steel with copper is inhibited by reaction with sulphur in the furnace gas. Considerations in the use of steels alloyed with copper are given. The author also describes works tests on a pusher furnace fired with coke-oven gas and oil.

Present State of Knowledge of Metal Fatigue. P. Lefort. (*Génie Civil*, 1956, **133**, Sept., 1, 318–321; Sept. 15, 334–336). A review of the effects of surface treatments and of alternating stresses on structure and properties in relation to fatigue.

International Conference on Fatigue of Metals. 1956. (*Sponsored by the Inst. Mech. Engrs. and A.S.M.E.*). **The Changing Nature of the Fatigue Problem.** H. J. Gough. (*Introductory paper*, pp. 9). A review of earlier conferences. **Allowance for Stress Concentrations in Design to Prevent Fatigue.** H. J. Grover. (*Session* 2, *Paper* 1, pp. 9). Tests under alternating and steady and alternating loads are given, with some consideration of structures. **Damping and Resonant Fatigue Behaviour of Materials.** B. J. Lazan. (*Session* 2, *paper* 2, pp. 14). The relation of damping to reversed stress is discussed. **The Effect of Mean Stress on the Push-Pull Fatigue Properties of an Alloy Steel.** H. C. O'Connor and J. L. M. Morrison. (*Session* 2, *paper* 2, pp. 10). The method of testing is described and the results in relation to the three static stress discontinuities are discussed. **A Method of Estimating the Fatigue Strength of Small Ellipsoidal Cavities.** R. E. Peterson. (*Session* 2, *paper* 4, pp. 10). With an Appendix on non-propagating cracks. **The Fatigue Testing of Components. A Means of Revealing the Danger Point in Pieces or Structures.** H. de Leiris. (*Session* 2, *paper* 5, pp. 6). **Fatigue Strength in Shear of an Alloy Steel.** W. T. Chodorowski. (*Session* 2, *paper* 6, pp. 12). Apparatus is described and results on a nickel–chromium–molybdenum steel, with special reference to the effect of mean stress and directional properties. **Further Results of Fatigue under Triaxial Stress.** J. S. C. Parry. (*Session* 2, *paper* 7, pp. 8). The fatigue limit for cylinders is greatly raised by honing the bore or by lining with rubber. **The Effect of Large Hydrostatic Pressures on the Torsional Fatigue Strength of an Alloy Steel.** B. Crossland. (*Session* 2, *paper* 8, pp. 14). Fluid pressure has no effect on the stress–strain curve except to increase the strain to failure by about 80% at 20 t/in². The deleterious effect of the oil may explain the low values referred to in the previous paper. **Theory of Combined Bending and Torsion Fatigue, with Data for SAE 4340 Steel.** W. N. Findley, J. J. Coleman, and B. C. Hanley. (*Session* 2, *paper* 9, pp. 10). S–N data are adequately represented by a 4–constant power function of N where one is the endurance limit. Five theories are considered and that of principal shear stress influenced by normal stress on the plane of failure and corrected for anistropy best represents the data. **Influence of Plastic Deformation on Notch Sensitivity in Fatigue.** P. G. Forrest. (*Session* 2, *paper* 11. pp. 15). Measurements are given on iron, steels, and other metals. **Interpretation of Fatigue Strengths for Combined Stresses.** J. Marin. (*Session* 2, *paper* 12, pp. 13). A theory is developed and compared with many published measurements. **Effect of Unintentional Stress Raisers on the Fatigue Strength of Structural Components.** E. C. Hartmann. (*Session* 2, *paper* 13, pp. 8). Accidental scratches are often of less importance than design features, but if they should coincide some effect may be produced. **Usefulness and Role of Repeated Strain Testing as an Aid to Engineering Design and Practice.** R. W. Bailey. (*Session* 2, *paper* 14, pp. 7). Hysteresis curves are given and interpreted. **Short Endurance Fatigue.** A. C. Low. (*Session* 2, *paper* 15, pp. 8). **Stress Concentration in Relation to Fatigue.** H. L. Cox. (*Session* 2, *paper* 16, pp. 12). Stress gradient and intrinsic flaws are discussed in relation to crankshafts in torsion. **Mechanical Aspect of Size Effect on Fatigue of Metals.** G. V. Uzhik. (*Session* 2, *paper* 17, pp. 6). **An Analysis of the Effects of Shot-peening upon the Fatigue Strength of Hardened and Tempered Spring Steel.** A. G. H. Coombs, F. Sherratt, and J. A. Pope. (*Session* 3, *paper* 1, pp. 10). **Cumulative Fatigue Damage.** H. T. Corten and T. J. Dolan. (*Session* 3, *paper* 2, pp. 14). Damage is visualized as nucleation of submicroscopic voids which develop into cracks. **The Influence of Fillet Radius on the Fatigue Strengths of Large Steel Shafts.** S. F. Dorey and G. P. Smedley. (*Session* 3, *paper* 3, pp. 12). **Distribution Functions for the Prediction of Fatigue Life and Fatigue Strength.** A. M. Freudenthal and E. J. Gumbel. (*Session* 3, *paper* 5, pp. 12). **Influence of Residual Stresses on the Fatigue Limit.** H. Sigwart. (*Session* 3, *paper* 6, pp. 12). **Fatigue of Curved Surfaces in Contact under Repeated Load Cycles.** N. G. Kennedy. (*Session* 3, *paper* 7, pp. 10). **Design of Screw Fastenings Subject to Repeated Stresses.** A. Erker. (*Session* 3, *paper* 8, pp. 13). **Random Noise Fatigue Testing.** A. K. Head and F. H. Hooke. (*Session* 3, *paper* 9, pp. 5). A machine and its calibration and use are described. **Effect of Variable Load and Cumulative Damage on Fatigue in Vehicle and Airplane Structures.** E. Gassner. (*Session* 3, *paper* 10, pp. 8). **On the Endurance of Cast Iron and Steel under Repeated Loading of Varying Amplitude.** S. W. Serensen. (*Session* 3, *paper* 11, pp. 9). **The Influence of Temperature on the Fatigue of Metals.** N. P. Allen and P. G. Forrest. (*Session* 4, *paper* 1, pp. 16). **Corrosion Fatigue.** A. J. Gould. (*Session* 4, *paper* 2, pp. 9). A comprehensive review of the phenomenon and factors involved in it is presented with notes on prevention. **An Investigation on the Corrosion Fatigue of Marine Propeller Shafts.** S. Hara. (*Session* 4, *paper* 3, pp. 6). **Fatigue of Large Shafts by Fretting Corrosion.** O. J. Horger. (*Session* 4, *paper* 4, pp. 11). **Very High-speed Fatigue Testing.** A. R. Wade and P. Grootenhuis. (*Session* 4, *paper* 5, pp. 11). A torsional vibrator is described giving absolute values for fatigue, measurements of dynamic modulus of elasticity and early detection of cracks. **Fatigue Properties of Steel at Higher Temperatures.** F. Wever. (*Session* 4, *paper* 6, pp. 7). Fatigue strength and mean stress, stress fracture times for static and alternate loading, notch effect and surface hardening, and metallography are discussed. **The Influence of Frequency of Vibration on the Endurance Limit of Ferrous Alloys at Speeds up to 150,000 cycles per Minute Using a Pneumatic Resonance System.** T. W. Lomas, J. O. Ward, J. R. Rait, and E. W. Colbeck. (*Session* 4, *paper* 7, pp. 13). **Fretting Corrosion and its Influence on Fatigue Failure.** A. J. Fenner, K. H. R. Wright, and J. Y. Mann. (*Session* 4, *paper* 8, pp. 10). **Corrosion Fatigue Cracking Resulting from Wetting of Heated Metal Surfaces.** A. H. Goodger. (*Session* 4, *paper* 9, pp. 9). The paper refers especially to steam power plant quoting failures of superheaters, boiler tubes and drums, desuperheaters, economizers etc. and discussing mechanism of attack and methods of detection. **The Effect of Fatigue Stresses on Creep and Recovery.** A. J. Kennedy. (*Session* 4, *paper* 10, pp. 9). A method of investigation with preliminary discussion is given. **Fatigue of Metals under Contact Friction.** I. A. Oding and V. S. Ivanova. (*Session* 4, *paper* 11, pp. 8). A continuous reduction of fatigue strength is observed which is attributed to a thermoelectric effect and which is prevented by a reverse current. Mechanism is discussed. **A Study of the Strain Cycling and Fatigue Behaviour of a Cold-worked Metal.** L. F. Coffin, jun. and J. H. Read. (*Session* 5, *paper* 1, pp. 12). Work on 347 stainless steel is described. Strain softening is discussed and explained in terms of annihilation of dislocations of opposite sign. **Relation of Inclusions to the Fatigue Properties of High-strength Steels.** F. B. Stulen, H. N. Cummings and W. C. Schulte. (*Session* 5, *paper* 4, pp. 8). After heat-treatment to ultimate tensile strength it appears that fatigue life depends largely on the size and location of inclusions. **The Influence of Some Surface Factors on the Torsional Fatigue Strength of Spring Steels.** J. F. Watkinson. (*Session* 5, *paper* 5, pp. 16). Shot peening is discussed. **Some Preliminary Fatigue Results on a Steel of up to 800 V.P.N. Hardness, using Notched and Unnotched Specimens.** J. E. Russell and D. V. Walker. (*Session* 5, *paper* 6, pp. 5). **Fatigue of Wrought High-Tensile Alloy Steels.** P. H. Frith. (*Session* 5, *paper* 7, pp. 41). 25 pp. of tables are given. **Hydrogen Occlusion and its Effect on the Fatigue Properties of Plain Carbon Spring Steels.** J. S. Jackson. (*Session* 5, *paper* 8, pp. 8). Occluded hydrogen has very little effect except in stress ranges above the fatigue limit. **The Fatigue Behaviour of Iron with Intergranular Weakness.** H. R. Tipler, and P. G. Forrest. (*Session* 5, *paper* 10, pp. 5). An unusual type of failure is described. **The Behaviour of Single Crystals of Iron under Fatigue Loading.** H. A. Lipsitt and G. T. Horne. (*Session* 6, *paper* 1, pp. 9). **Studies in the Formation and Propagation of Cracks in Fatigue Specimens.** N. E. Frost and C. E. Phillips. (*Session* 6, *paper* 2, pp. 9). **Experiments Relating to the Basic Mechanism of Fatigue.** N. Thompson. (*Session* 6, *paper* 3, pp. 6). Work on copper and nickel confirms previous findings on

aluminium and mild steel. **Failure of Metals under Cyclic Strain.** W. A. Wood. (*Session* 6, *paper* 4, pp. 6). **The Basic Mechanism of Fatigue and its Dependence on the Initial State of a Material.** P. J. E. Forsyth. (*Session* 6, *paper* 5, pp. 5). Microscopical changes preceding fatigue damage are described. **The Growth of Fatigue Cracks under Plastic Torsion.** F. A. McClintock. (*Session* 6, *paper* 6, pp. 7). Theory is outlined and preliminary measurements given. **Metallographic Observations on the Fatigue of Steels.** M. Hempel. (*Session* 6, *paper* 7, pp. 7 + 15 plates). **The Distinction between Initiation and Propagation of a Fatigue Crack.** J. A. Bennett. (*Session* 6, *paper* 8, pp. 5). **Strength Reduction Factors for Small Quenching Cracks and for Decarburized Steel.** J. A. Pope and C. W. Barson. (*Session* 6, *paper* 10, pp. 8). **Lessons to be Learnt from Failures in Service.** G. A. Cottell. (*Session* 7, *paper* 1, pp. 9). Dry-fatigue and corrosion fatigue in service are discussed and general principles for their prevention are stated. **Fatigue in Automobiles.** R. J. Love. (*Session* 7, *paper* 2, pp. 13). **Fatigue Failure and Service Experience with Particular Reference to the Shape of the Part.** R. Cazaud. (*Session* 7, *paper* 3, pp. 6). Keyways and other features of shafts are considered. **Fatigue of Plain Bearings.** W. E. Duckworth and G. H. Walter. (*Session* 7, *paper* 4, pp. 10). **Fatigue, Residual Stress and Surface Cold Working.** R. L. Mattson. (*Session* 7, *paper* 5, pp. 13). **Fatigue Testing of Compression-type Coil Springs.** R. C. Coates and J. A. Pope. (*Session* 7, *paper* 6, pp. 15). Apparatus is described and a mathematical treatment given, with discussion of the effects of shot-peening scragging, and both. **Shot-peening as Protection of Surface-damaged Propeller-blade Materials.** J. M. Lessells and R. F. Brodrick. (*Session* 8, *paper* 1, pp. 13). **Fatigue in Aero-Engines.** E. R. Gadd. (*Session* 8, *paper* 5, pp. 16). **Fatigue in Railroad Equipment.** W. M. Keller and G. M. Magee. (*Session* 9, *paper* 1, pp. 5). **Fatigue Tests on Thin-gauge Box-section Beams.** R. P. Newman and G. Coates. (*Session* 9, *paper* 2, pp. 9). **Fatigue Tests on Seamless Mild-Steel Pipe Bends.** P. H. R. Lane. (*Session* 9, *paper* 3, pp. 10). **Significance of the Fatigue of Metals to Railways.** T. Baldwin. (*Session* 9, *paper* 4, pp. 11). **The Fatigue Problem in Welded Construction.** R. Weck. (*Session* 9, *paper* 5, pp. 11). **Service Fatigue Failures in Marine Machinery.** T. W. Bunyan. (*Session* 9, *paper* 6, pp. 12 + 10 plates). **Some Studies on the Pitting of Marine Reduction Gears.** J. Hoskins. (*Session* 9, *paper* 7, pp. 6).

Failure under Repeated Loads. S. I. Ratner. (*Doklady Akad. Nauk. SSSR.*, 1956, **106,** (2), 246–249). Four steels were used and notched test-pieces were examined. Three topics are discussed: whether tensile strength or resistance to brittle fracture is more important, whether brittle fracture resistance changes and how it varies with the character of the material, and what is the effect of time during repeated loading tests.

On the Frequency Dependence of the Fatigue of Metals. T. Yokobori. (*J. Phys. Soc. Japan*, 1956, **11,** June, 715–716). An attempt is made to explain the discrepancy between the author's theory of proportionality between stress cycle frequency and number of cycles to fracture in fatigue, and data reported by Head for non-ferrous metals.—J. G. W.

A Rapid Method of Estimating the Fatigue Limit or Endurance Limit of Metals in Reverse Bending. J. McKeown. (*Metallurgia*, 1956, **54,** Sept., 151–156, 158). An account of the application of a progressive loading technique to the determination of fatigue or endurance limit is presented. It is suggested that the method might find its most important application in determining the effect of changes in design in improving or detracting from fatigue performance.—B. G. B.

Contribution to the Study of the Alterations in Steel Structures Caused by Fatigue. A. Kochanovscká, J. Čermák and F. Holý. (*Rev. Mét.*, 1956, **53,** Sept., 701–702). Studies on low alloy Ni–Cr–Mo and Mn–Si–Cr steels using the change in X-ray pattern to measure the development of fatigue are reported.—B. G. B.

A Method of Predicting the Effects of Notches in Uniaxial Fatigue. W. E. Dirkes. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78,** April, 511–515). A semi-graphical method of extrapolating notched test data is presented by which the notched fatigue strength of materials can be predicted if the unnotched fatigue strength is known for various stress ratios. The following minimum notched test data are required. (*a*) the fatigue strength with a static load greater than the peak alternating load and (*b*) the static tensile strength. The analysis is in reasonable agreement with available test data for Al and steel.—D. H.

The Measurement of Fatigue Damage in Mild Steel. P. G. Forrest. (*Engineering*, 1956, **182,** Aug. 31, 266–268). Test pieces were subjected to fatigue stress and examined for progressive damage and rate of crack propagation by tensile testing at − 196° C. The method proved rather insensitive, due to the ductility still present after fatigue.

A Criterion on Fatigue Failure of Metals. Y. Kawada and S. Kodama. (*Mem. Fac. Tech. Tokyo Univ.*, 1956, No. 6, 321–329). [In English]. A new mechanism for fatigue failure is proposed. The metal is fatigued by repeated shear stress, and its strength is reduced in proportion to the stress amplitude. Failure is caused by normal or shearing stresses; these are separating or sliding failures, respectively. Failure occurs in the plane where the combined effects are at a maximum. This mechanism adequately explains the compression fatigue limit diagram. (17 references).—K. E. J.

Scatter of Fatigue Life of Structural Steel and its Influence on Safety of Structure. I. Konishi and M. Shinozuka. (*Mem. Fac. Eng. Kyoto Univ.*, 1956, **18,** Apr., 73–83). [In English]. Fatigue tests were made on a structural steel; the distribution can be interpreted on a stochastic basis. Diagrams are given for estimating the probability of non-failure under constant stress or when two kinds of repeated stress are applied. The relationship between stress level and probability of failure is linear on the logarithmic scale: this allows the probability of non-failure under varying stress to be estimated.—K. E. J.

A Wire Fatigue Machine for Investigation of the Influence of Complex Stress Histories. H. T. Corten and G. M. Sinclair. (*Amer. Soc. Test. Mat. Preprint No.* 67, 1956). A wire fatigue testing machine is described that operates on the principle of a deflected rotating strut. This equipment was designed to investigate fatigue life for complex stress histories, particularly those consisting of repeated blocks of cycles. With this wire machine the stress amplitude may be changed rapidly from one level to another without inducing noticeable vibration in the wire specimen. The results of constant stress amplitude tests of steel wire are presented and analysed statistically.—B. G. B.

Resistance of Steels to Vibration at Elevated Temperatures. I. Conclusions from Hot Vibration Tests. F. Wever and M. Hempel. (*Forschungsberichte des Wirtschafts- u. Verkehrsministeriums Nordrhein—Westfalen*, 1956, (311), pp, 36). **II. Push–Pull–Vibration Tests on Two Heat-resisting Steels at 500–650°** (1956, (312), pp. 36). Creep tests and fatigue studies on Armco iron and Nimonic alloys and steels are described at high temperatures with some studies on corrosion effects and hardness measurements. Chrome and niobium steels are included. In part II the apparatus and form of test pieces are described and tests are applied to a ferritic Cr–Mo–V and to an austenitic Cr–N–Mo–Nb steel and microstructures are reproduced. Findings are discussed.

Design Aspects of High-Temperature Fatigue with Particular Reference to Thermal Stresses. L. F. Coffin. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78,** April, 527–532). A criterion for fatigue failure has been proposed, based on experiments carried out on test specimens subjected to constrained thermal cycling and constant-temperature strain cycling. This criterion relates the number of cycles to failure with the plastic strain changes per cycle. The application of such a criterion to design where thermal-stress fatigue is the principal factor is discussed from fundamental and practical viewpoints. From such a criterion it is possible to predict the life of a machine part for a calculated thermal stress or conversely, the thermal stress permitted for a certain limiting number of stress cycles.—D. H.

Influence of Surface Effect on the Initiation of Fatigue Cracks. O. Lissner. (*Jernkontorets Ann.*, 1956, **140,** (5), 360–372). [In Swedish]. Fatigue tests were made on 3 carbon steels and one quenched and tempered NiCr steel in which a surface effect was observed to contribute to fatigue cracking. A very thin surface layer was removed after half the average life of the sample had elapsed, whereupon the test was continued. In all cases a considerable increase in specimen life was observed. In the carbon steels ageing was found to contribute to this effect. The results obtained support Schaub's theory that the fatigue mechanism involves not only plastic deformation but also chemical re-

actions, especially between atmospheric oxygen and the atoms in the surface zone of the metal.—G. G. K.

The Stress Rupture Strength of Type 347 Stainless Steel Under Cyclic Temperature. E. E. Baldwin. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78,** April, 517–525). Stress-rupture tests of 18/10/NbTa stainless steel were conducted in liquid sodium under constant and cyclic temperature conditions. Constant temperature tests were conducted at 1000–1200° F and cyclic-test temperatures of 416–1294° F and times 6–12 h. It was concluded that the deviation of the test results from the rupture life calculated by the method of Robinson and Miller was due to transient creep under cyclic-temperature changes.—D. H.

Lead Additions: Effect on the Fatigue Properties of En. 24 and En. 36 Steels. W. E. Bardgett. (*Iron Steel*, 1956, **29,** Aug., 392–395). By means of microscopical and microradiograph examination, mechanical testing, and fatigue testing, the author has studied the effect of additions of 0·2% Pb on the fatigue properties of 1½% Ni–Cr–Mo (En. 24) direct-hardening steel heat-treated to three tensile strengths, and of 3% Ni–Cr (En. 36) case-hardening steel heat-treated to two tensile levels. The results indicate that the lead addition only lowers the rotary bending fatigue strength substantially at high tensile strength levels and where no stress raiser is present, conditions rarely arising in engineering practice.

Residual Stresses After Heat Treatment of Unstable Manganese Steels. H. Bühler and E. Herrmann. (*Arch. Eisenhüttenwesen*, 1956, **27,** Apr., 269–274). Investigations on hard manganese steel containing 1·2% C and 13·6% Mn are described. The heat-treated forms used were: water-quenched from 1050° C, as rolled, annealed up to 700° C, and bath-hardened.

Influence of Cold Work on Strength of Steel at Elevated Temperatures. P. Shahinian. (*Trans. Amer. Soc. Metals.*, 1956, **48,** 952–970). The influence of cold work on creep, rupture and relaxation properties of a quenched and tempered Cr–Mo steel was investigated. The results are explained in terms of the Bauschinger effect.—E. E. W.

An Analysis of the Phenomenon of High-Temperature Creep. O. D. Sherby and J. E. Dorn. (PB 115924, 1954, pp. 42, *T.I.D.U. List* 571). A review is presented of known data on high-temperature creep. It is shown that the activation energy for creep for a given metal is insensitive to dilute solid solution alloying, grain size, dispersions of stable intermetallic compounds, cold-working, temperature stress and structure. Energies of activation for creep of pure metals are approximately equal to the known values of activation energies for self-diffusion. Creep-rate dependence of metals on the applied stress for constant structure is reviewed. A model for creep is explained. It suggests that the rate controlling process is that of recovery of barriers from the path of moving dislocations and that the removal of these barriers is controlled by atomic diffusion processes.

Study on High-temperature Creep and Rupture of Heat Resisting Steel, II. T. Mishima, F. Nakanishi, T. Akutagawa, J. Miyama, K. Hatta, and T. Okazaki. (*Tokyo Univ. Fac. Eng. Annual Report*, 1954, 19–23, summary, 243). An improved testing machine is described and 18–8 steels have been tested at 600°, 650°, 670°, and 700° C at 2·8–8 Kg/mm² for 600 h. An equation $\epsilon - \epsilon_0 = At^b$ is followed where ϵ is total and ϵ_0 is initial plastic deformations, t is time and A and b are constants.

Effect of Stress or Creep at High Temperatures. H. Laks, C. D. Wiseman, O. D. Sherby, and J. E. Dorn. (PB 118377, 1954, pp. 28, *T.I.D.U. List* No. 571). Experimental investigations revealed that the high-temperature creep rate ϵ is related to the stress δ by $\epsilon \sim \delta^n$, for low stresses and by $\epsilon \sim e^{B\delta}$ for high stresses where n and B are constants which are independent of creep strain and temperature. According to a preliminary dislocation climb model for high temperature creep, the activation energy for creep should be that for self-diffusion, and the effect of stress on creep rate should depend on the number of active Frank-Read sources, while the rate of climb depends on the structure as determined by the pattern of climbing dislocations. Many of the experimental observations on high-temperature creep can be explained by this model.

Recording Temperature During a Creep Test. (*Machinery*, 1956, **89,** Oct., 19, 906). The Servograph Mark 2 recorder is shown in use.

Effect of Alloying Elements upon the Creep Strength of Ferritic Heat-resisting Steels (1st Report). T. Akutagawa and T. Fujita. (*Tokyo Univ. Fac. Eng. Annual Report*, 1955, 133–138). Steel with 12% Cr was studied for the effects of heat treatment and for the addition of Mo, W, V, Nb, and Ti. Specimens were heated for 30 min at 1150° or at 1000° C, oil quenched and annealed 1 h at 680°, cooling in air. Creep tests at 620° were carried out. Creep strength was increased by all additions but Nb; simultaneous addition of several elements gave increases greater than the sum of the individual additions.

Corrigendum

In the *Journal*, Sept. 1956, p. 105, Col. 1, 7th abstract. "**Changes in the Structural Constitution of Chrome–Nickel–Molybdenum Steels under Sustained Load in Creep Tests at 500°**" the authors should read: "F. Wever, A. Krisch and H.-J. Wiester," the first name having been omitted.

On the Periodic Creep of Heat Resisting Alloys. G. Vidal. (*Rev. Mét.*, 1956, **53,** July, 485–496). The action of a periodic stress may, in certain circumstances, be a cause of rupture due to periodical creep, but not to real fatigue. Three diagrams have been constructed which allow the calculation from creep data at constant stress of the constant effective stress the effect of which is in theory the same as that of a given periodical stress. The three diagrams are respectively relative to stresses varying in time, according to sinusoidal, rectangular or triangular laws. The results of tests carried out at 800° C on a Nimonic 90 alloy confirm the validity of the theoretical diagrams.—B. G. B.

Creep in Metal Crystals at Very Low Temperatures. N. F Mott. (*Phil. Mag.*, 1956, **1,** (8th series), June, 568–572) Theoretical.

Creep Stresses and Deflections of Columns. T. H. Lin. (*J. Appl. Mech.*, 1956, **23,** June, 214–218). A direct method of calculating the stresses and deflection of a pin-ended column under creep is described.—D. H.

Creep and Creep Rupture Characteristics of some Rivetted and Spot Welded Lap Joints of Aircraft Materials. L. Mordfin. (*NACA–TN*3412, 1955, pp. 54; *T.I.D.U. List* 443). Rivetted aluminium and spot-welded stainless steel joints were studied, the latter at 800° F. With bibliography.

Analysis of Creep in Rotating Disks Based on the Tresca Criterion and Associated Flow Rule. A. M. Wahl. (*J. Appl. Mech.*, 1956, **23,** June, 231–238). Mathematical treatments of discs with (*a*) constant thickness and constant temperature, (*b*), variable thickness and constant temperature, and (*c*) variable thickness and variable temperature are given. The starting period is treated on various assumptions.

On the Theory of the Low-Temperature Internal Friction Peak Observed in Metals. A. Seeger. (*Phil. Mag.*, 1956, **1,** 8th Series, July, 651–662).

Fundamental Causes of the Destruction of Teeth. A. R. Purdy. (*Rev. Univ. Min.*, 1956, 9th series, **12,** Sept., 382–392). A review of the fundamental causes of failure of gear teeth is presented in order to diagnose the reason for their failure under service conditions. 27 photographs are reproduced which show typical structures of broken gear teeth.

Brinell Hardness Testing. J. Léonard. (*Found. Trade J.*, 1956, **101,** Aug. 16, 173–175). The results of comparative hardness tests on cast irons using two combinations of ball diameter and load, i.e. 2·5 mm/187·5 kg and 10 mm/3000 kg respectively, are reported and discussed. The 2·5 mm/187·5 kg combination is preferred for machineable grey iron castings as it is claimed that it permits accurate reading, testing close to an edge and closer spacing of the impressions.

A Note on the Measurement of Vickers Hardness Indentations. B. S. Satyanarayana. (*Canad. J. Techn.*, 1956, **34,** Sept., 375–377). An interferometric study of the shape of indentations is reported and errors are evaluated from the distortion found.

Effect of Prolonged Heating at High Temperature on the Hardenability of Boron-Treated Steels. R. M. Goldhoff, R. Speiser and J. W. Spretnak. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206;** *J. Met.*, 1956, **8,** Aug., Section 1, 1091). Previous work by Grange and Garvey indicated that homogenization of boron-treated steels could lead to complete elimination of the boron hardenability effect. The authors have studied this by prolonged heating at high temperatures

of Grainal-treated steels, and conclude that no loss of boron hardenability effect occurs when deboronization is circumvented.—G. F.

Nature and Applications of Microhardness. H. Bückle. (*Metall*, 1955, **9**, July, 549–554; Dec., 1067–1074). A survey describing microhardness measurement technique with emphasis on errors encountered, and summarizing data on the effects of geometrical and metallurgical variables. In the second part Vickers hardness is considered and then, in some detail, the uses of the method for the investigation of metallographic features, and the testing of hardened parts and non-uniform profiles.

Microhardness Penetration Curves in Two Nickel-bearing Carburized Steels. H. M. Otte. (*J. Iron Steel Inst.*, 1957, **186**, May, 13–21). [This issue.]

Hardenability Factors for Hypereutectoid Low-Alloy Steels. E. J. Whittenberger, R. R. Burt, and D. J. Carney. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 2, 1008–1016). Hardenability factors for the elements, C, Mn, Si, Cr, Ni, and Mo, have been developed for use with hypereutectoid low-alloy steels in which bainite is the first sub-critical transformation product. They permit the prediction of 95, 80, and 50% martensite hardenability when quenching previously spheroidized steels from 1475°, 1525°, and 1575° F. The strong contributions of Si and Mn suggest their use to obtain additional hardenability in this type of steel.—G. F.

Aptitude to Ageing of Various Steels shown by Simple Brinell Tests. E. Houdremont, W. Wepner and H. J. Wiester. (*Rev. Mét.*, 1956, **53**, Oct. 750–756). A Brinell hardness test carried out between room temperature and 300° C under standard conditions enables various steels to be distinguished according to their sensitiveness to ageing in a simple and rapid manner. It is possible to show the effects of nitrogen, carbon, deoxidation by aluminium and heat treatment on the age-hardening properties of steels.—B. G. B.

Strain Ageing of AISI 4340. H. W. Paxton and C. C. Busby. (*Trans. Amer. Inst. Min. Met., Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, June, 788–789). The authors have studied the effects of strain ageing, following tension or rolling, on the tensile properties of low-carbon martensitic AISI 4340 steel. The treatment following rolling was found to be consistent in producing a tensile strength of 260,000 lb/in² with a yield-tensile ratio of about 0·95.—G. F.

Reflections on the Question of Brittle Fracture. E. Houdremont and H. Mussmann. (*Stahl u. Eisen*, 1956, **76**, July 12, 903–907). A critical review is given of the various methods used in determining the temperature of transition from ductile to brittle fracture. The significance of the energy taken up in fracture is considered, and the nature of the curves of notch impact strength against temperature is discussed.

Notes on the Brittle Fracture of Steels. W. Felix and T. Geiger. (*Sulzer Tech. Rev.*, 1956, **38**, (1), 14–27). Mechanism is investigated in the high and low ranges of notch impact values. X-ray photographs are shown and interpreted. The destruction of an I-girder is then examined chemically, metallographically, mechanically, by X-rays and by microhardness measurements. The effect of the position of crystals is demonstrated though without quantitative information on the dependance of deformation of ferrite on orientation.

On Brittle Failure in Cast Steels. O. V. Stupishina and V. I. Likhtman. (*Doklady Akad. Nauk SSSR.*, 1956, **107**, (2), 252–254). Tensile strength-hardness curves are shown on samples heated for short periods at temperatures between 160–700° C.

The Influence of Molybdenum and Tungsten on Temper Embrittlement. A. E. Powers. (*Trans. Amer. Soc. Metals*, 1956, **48**, 149–164). The influence of up to 2% W and Mo on the susceptibility of a 1% Cr, 1% Mn steel has been studied. The rise in transition temperatures was used to measure the susceptibility to embrittlement of the steels which were aged for 1000 hours at various temperatures within the temper embrittling region.—E. E. W.

Embrittlement of a Ship Steel in a Nitrate Salt Bath. D. J. Garibotti, C. A. Beiser, and E. B. Evans. (*Welding J.*, 1956, **35**, Aug., 383s–388s). An investigation was made to determine the cause of the severe embrittlement of a 1025 semi-killed ship steel (C steel) after sub-critical heat treatment for long times in a nitrate salt bath. The ductility was evaluated by means of Charpy **V**-notch transition temperatures. Metallographic studies, X-rays and chemical analyses showed that the embrittling agent was nitrogen introduced by a scalding reaction of the steel with the nitrate.—U. E.

Wear Resistance: Recent Russian Research on Different Surface Treatments. W. G. Cass. (*Iron Steel*, 1956, **29**, July, 342). A summary is given of recent Russian research by Kazartsey on the relative wear resistance of surface coatings used in the reconditioning of machine parts. The treatments compared include electro-fusing or build-up, chrome-plating, iron-plating, metallizing, and electro-sparking erosion.—G. F.

Resistance of Metals and Minerals, Including Very Hard Materials, to Abrasive Wear. M. M. Khrushchov and M. A. Babichev. (*Doklady Akad. Nauk SSSR.*, 1956, **107**, (1), 75–78. Pressures exceeding 2000 Kg/mm² were used on a variety of non-ferrous metals and carbides.

Residual Strain after Abrasion of Metals. O. G. Karpinskiĭ and B. M. Levitskiĭ. (*Doklady Akad. Nauk SSSR.*, 1956, **106**, (1), 55–57). Armco iron specimens were ground parallel with a special rotary apparatus using an abrasive paper and examined with cobalt Kα radiation. Strain is related to abrasion rate in m per min.

Cast Iron in Diesel Engines. J. G. Pearce. (*Sci. Lubrication*, 1956, **8**, Mar., 28–30). Wear resistance of liners and piston rings is reviewed.

The Relationship Between Static and Kinetic Friction. N. F. Kunin and G. D. Lomakin. (*Zhur. Tekhn. Fiz.*, 1954, **24**, [8], 1367–1370). The processes are described in terms of contact area and plastic flow and formulated mathematically Experiments are described with tin on cast iron, and rupture and strain hardening effects were studied with slow increases of tangential force. The ratio is expressed in terms of a constant γ which is a coefficient of frictional strain-hardening. The measurements agree with the ordered microrupture effect put forward.

Measurements by Means of the Versatile Perth-O-Meter Surface-Finish Testing and Measurement. Dr. J. Perthen's System. (*Microtecnic*, 1956, **10**, (3), 149–151). A description is given of the Perth-O-Meter surface testing and measuring instrument, which employs a system of interchangeable tracers which can be adapted to various forms of workpiece and to different problems of measurement. The instrument is intended for workshop use.—L. D. H.

Measuring Surface Finish by Electronics. S. F. Smith. (*Indust. Finishing*, 1955, **9**, Dec., 274–278). Stylus methods for measurement of surface finish are described and systems of measurement are discussed.—A. D. H.

Magnetism in Retrospect and Prospect. E. C. Stoner. (*Year Book of Physical Society*, 1955, 23–43). The 39th Guthrie Lecture.

Magnetization Curves and Domain Structure. L. F. Bates and A. Hart. (*Proc. Phys. Soc.*, 1956, **69B**, May, 497–505). Measurements on a silicon-iron crystal are described and treated theoretically.

Theory of Neutron Scattering by Conduction Electrons in a Metal and on the Collective-electron Model of a Ferromagnet R. J. Elliott. (*Proc. Roy. Soc.*, 1956, **235A**, May 8, 289–304).

Spin Fluctuation Scattering of Neutrons and the Ferromagnetic State of Iron. R. D. Lowde. (*Proc. Roy. Soc.* 1956, **235A**, May 8, 305–320).

Ferromagnetic Domain Nucleation in Silicon Iron. L. F. Bates and D. H. Martin. (*Proc. Phys. Soc.*, 1956, **69B**, Feb., 145–152). Observations with 3% silicon iron crystals are consistent with domain phase creation at surface inclusions.

The Hall Effect in Pure Iron and in Steel Containing 1·2% C. W. Jellinghaus and M. P. de Andrès. (*Arch. Eisenhüttenwesen*, 1956, **27**, Mar., 187–192). The fundamentals of the Hall effect in ferromagnetic materials are considered. The determination of the Hall constants for pure iron, Armco iron, and for 1·2% C and 4% C steels is described, and the results are compared with previous measurements. Changes after heat treatment of hardened steels are discussed.

Hall Effect and Magnetic Properties of Armco Iron. S. Foner. (*Phys. Rev.*, 1956, **101**, March 15, 1648–1652). The Hall effect, which was shown to be useful for studying detailed behaviour of magnetic properties, was found to be the sum of the ordinary and the extraordinary effects over the range investigated, including the non-linear region. The extraordinary constant was independent of the field. The approach to magnetic saturation was examined at high fields.

A Non-Destructive Method for Measuring the Thickness of Metallic Surface Layers on Metals. T. F. Alkér and E. Schalk. (*Acta Techn.*, 1956, **15**, (1–2), 3–18). [In English].

Problems involved in the design of thickness meters are reviewed. A high-frequency magnetic induction bridge was developed and tested on hardened surface layers of steels. The construction, characteristics, and design criteria of the equipment are given, together with results obtained on nitrided and carburized plain and alloy steels.—P. F.

The Density, Magnetic Properties, Young's Modulus, and ΔE Effect, and their Changes due to Quenching in Ferro-Magnetic Iron–Aluminium Alloys. I. The Density and Magnetic Properties. M. Yamamoto and S. Taniguchi. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, A., 1956, **8,** Apr., 112–124). Measurements of density and ferromagnetic properties of annealed and quenched iron–aluminium alloys containing less than 17% Al are reported. Density changes were in good agreement with calculations based on change in lattice parameter.—J. G. W.

Preferred Orientations and Magnetic Properties of Rolled and Annealed Permanent Magnet Alloys. W. R. Hibbard, jun. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8,** Aug., Section 2, 962–967). Following pole figure determination, torque tests, metallographic examination, and coercive force measurement, the author documents the rolling and annealing tectures of the following ductile permanent magnet alloys: Cunife, Cunico, Silmanal, Vicalloy, and Heusler's Alloy.—G. F.

Effect of Small Cold Reduction on the Structure and Magnetic Properties of Hot-Rolled Transformer Steel. B. F. Trakhtenberg. (*Stal'*, 1956, (4), 343–347). [In Russian]. It has been found that after cold reductions of 4–8% of transformer steel, the total electrical losses decrease by about 0·2–0·3 watt/kg. Through decreases in the grain size, the magnetic induction in strong fields decreases by 1–3%.—S. K.

The Temperature-dependent and Temperature-independent Parts of the Increase of Resistance due to Impurities in a Metal. A Simple Method of Obtaining their Ratio. A. N. Gerritsen and J. M. L. C. Van der Aa. (*Appl. Sci. Res.*, 1956, *A*, **6,** (2–3), 191–196).

On the Molecular Theory of Electrostriction. B. K. P. Scaife. (*Proc. Phys. Soc.*, 1956, **69B,** Feb., 153–160).

Progress in Ultrasonics—I. Generating Methods and Commercial Development. A. E. Crawford. (*Engineer Foundryman*, 1956, **21,** July, 46–51). Methods of high-power ultrasonic generation are classified and the characteristics of each method are discussed with reference to potential and actual industrial applications.—A. D. H.

Progress in Ultrasonics—II. Practical Applications. A. E. Crawford. (*Engineer Foundryman*, 1956, **21,** Aug., 48–53). The physical action of ultrasonics is described and examples of their application including precipitation of fumes and smokes and degreasing. Metallurgical effects such as refinement of grain size of castings, in soldering and in drilling of hard materials are described.—A. D. H.

The State of Ultrasonic Testing in Germany and the Application of this Method by the German Federal Railways. E. Martin. (*Nondestructive Test.*, 1956, **14,** July-Aug., 26–31). The application of ultrasonic testing to the inspection of axles on rolling stock and rails is described. Methods of automatic recording are presented.—A. H. M.

The Detection of Internal Flaws in Rails by Ultrasonic Methods. J. Palmé. (*Inspection Engineer*, 1956, **20,** July-Aug., 86–89). The author gives a brief description of the 'Sonirail' and the principles of its operation. The results obtained with this equipment during 2 years service on the S.N.C.F. track are given.—A. H. M.

Ultrasonic Thickness Measurement. J. Carmichael. (*Inst. Petroleum Rev.*, 1956, **10,** Oct., 276–278, 280). A transportable transducer unit is described.

Ultrasonic Testing of Meehanite Castings. (*Instrument Practice*, 1956, **10,** Sept., 809–810). A modified ultrasonic flaw detection technique has been developed at Messrs. Goulds Foundries, Newport, for the testing of batches of Meehanite metal castings. A modified combined transceiver probe is used, and it is claimed that costs can be cut and production increased by the use of the technique.

Ultrasonic Gauging in Ship Repair Yards. (*Instrument Practice*, 1956, **10,** Sept., 808–809). A description is given of the application of an ultrasonic thickness gauge to the measurement of the thickness of the hulls of tankers, as an alternative to drilling holes. The expense of cleaning spots in the accessible surface and taking the measurements is appreciably less than the drilling technique, and a substantial amount of time is saved.—L. D. H.

Checking the Wall Thickness of Hollow Steel Shafts by the Multiple-back-reflection Immersion Method. J. Koziarski. (*Nondestructive Test.*, 1956, **14,** July-Aug., 12–15). The development of an immersion unit employing light hydraulic oil for checking the minimum wall thickness of helicopter-rotor shafts is described. Results were within 1·4%, while the contact ultrasonic method yielded results within 2·4%. Calibration procedure is given and the advantages of the method are discussed.—A. H. M.

Influence of Geometry upon an Ultrasonic Defect-size Determination in Large Rotor Forgings. S. Serabian. (*Nondestructive Test.*, 1956, **14,** July-Aug., 18–21). The author shows how the defect orientation and the geometrical configuration of the test forging dominate attempts to correlate the magnitude of ultrasonic indications and the defect sizes involved. Methods are presented to measure the angle at which the defect is lodged, along with the correction factors to be applied to eliminate this effect. The effect of forging geometry upon the testing sensitivity is discussed.—A. H. M.

Study of the Detection of Internal Flaws of Mill Rolls by Means of Ultrasonic Waves. M. Oyama and Y. Nozaka. (*Tokyo Univ. Fac. Eng. Annual Report*, 1954, 77–80, summary, 249). Model experiments are described and the best frequency and transducer diameter are derived.

Trends in Industrial Radiology. F. Lefay. (*Genie Civil*, 1956, **133,** Oct. 15, 369–373). Welding tests are discussed, among other topics. The use of an amplifier is particularly described as an aid to continuous observation.

Portable Projector Simplifies Radiographic Inspection. C. M. Dick. (*Foundry*, 1956, **84,** Sept., 228, 230, 233, 234). A description is given of the portable self-contained γ-radiography unit developed by the M. W. Kellogg Co. The sources used are ^{60}Co, ^{137}Cs, and ^{192}Ir. The smallest unit weighs 75 lb and is suitable for radiographing steel sections up to 2 in. thick.—B. C. W.

A High-temperature X-Ray Study on High-speed Steel. I—The Lattice Expansion of Matrix and Carbide. II—Transformations during Tempering and Some Considerations on Martensite-type Reactions. H. J. Goldschmidt. (*J. Iron Steel Inst.*, 1957, **186,** May, 68–85). [This issue.]

Study of Surface Films of Means of X-ray Fluorescence. C. Legrand and J.-J. Trillat. (*Rev. Mét.*, 1956, **53,** Aug., 645–648). The emission of characteristic rays from a substance submitted to X-rays of a suitable wave-length is used to measure the thickness of thin films. When the support for the film is not excited the intensity of the fluorescence due to the thin film increases linearly with its thickness. In other cases the fluorescence of the film support can be excited and the absorption of the emitted rays by the surface film can be used to measure its thickness.—B. G. B.

Non-destructive Testing of Metals By Gamma-Radiography. A. Blondel. (*Fonderie*, 1956, June, 213–222). The fundamentals of γ-radiography are reviewed. Among the aspects discussed are the available sources, the absorption of γ-rays, the effect of γ-rays on photographic emulsions, the factors influencing contrast and definition and the sensitivity of the technique. The practical aspects are briefly considered.

Material Testing of Stainless Steel by Means of Electron Diffraction. S. Yamaguchi. (*Z. Metallkunde*, 1956, **47,** Feb., 95–96). It is known that the surface of an austenitic stainless steel (18% Cr, 8% Ni) can be transformed to ferrite by mechanical polishing (cf. *Trans. AIME*, 1939, **135,** 486). The article describes the detection of this surface transformation by means of electron diffraction studies on wedge-shaped test pieces. The ferrite layer can be removed by electrolytic polishing.—L. D. H.

The Electron Spectrograph for Metal—Physical Investigations. W. Dietrich, H. Düker, and G. Möllenstedt. (*Z. Metallkunde*, 1956, **47,** Apr., 240–242). A description is given of the construction and operation of an electrostatic analyser of high resolving power, suitable for investigating the velocity spectrum of 50 Kv electrons.—L. D. H.

Improvements in High-sensitivity Fluoroscopic Technique. E. L. Criscuolo and D. Polansky. (*Non-Destructive Test.*, 1956, **14,** Jan-Feb., 30–31, 40). The performance of a high-sensitivity fluoroscope with a new high-power 4000-kvp-ma industrial fluoroscopic tube is described.—A. H. M.

Continuous Recording of Transformation and Cooling Curves. Description of the Dilatometer Employed. A.

Constant. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1956, **13**, No. 8, 1625–1635). A description of a dilatometer constructed by I.R.S.I.D. is given together with typical transformation and cooling curves obtained with an austenitic steel.—B. G. B.

The Thermal Expansion of Pure Metals and the Possibility of Negative Coefficients of Volume Expansion. J. H. O. Varley. (*Proc. Roy. Soc.*, 1956, **237**, Nov. 6, 413–421). Mechanism is advanced for the negative coefficient observed in plutonium.

Variation of the Viscosity of Metallic Wires with Temperature. C. Sălceanu. (*Compt. Rend.*, 1955, **241**, Nov. 14, 1384–1386). The results are given of experiments carried out on wires of seven different metals. For iron, the viscosity has a minimum value of $0{\cdot}76 \times 10^{-8}$ poises at 287° C.

A Study of Mechanical Properties at Very High Temperatures. C. Boulanger and C. Crussard. (*Rev. Mét.*, 1956, **53**, Sept., 715–728). Apparatus is described which records load–deformation curves from room temperature to 1500° C. An alternating torsion is applied to a 4–5 mm dia. sample. Creep or relaxation measurements up to 1 kgm/mm² are also possible together with the free damping of the sample under a 10 cycle/sec frequency. The machine has been designed to study the movement of dislocations at high temperature and results obtained are given and discussed.—B. G. B.

The Breakdown of Metals at High Temperatures. I. Ya. Dekhtyar and K. A. Osipov. (*Doklady Akad. Nauk SSSR.*, 1955, **104**, (2), 229–232). Equations are derived for changes as the melting point is approached. A quasi-liquid state seems to be formed before the structure collapses.

Changes in the Structure in Ferritic and Austenitic Special Steels by Prolonged Stress at High Temperatures. A. Schrader. (*Rev. Univ. Min.*, 1956, 9th series, **12**, Oct., 537–543). Electron microscopical examinations of high temperature steels submitted to creep tests with and without the application of external loads are described. The structural modifications can easily be followed if the chemical etch is such that precipitated carbides remain in the replica. By electrolytic isolation the composition of carbides can be determined allowing the structural modifications during high temperature creep tests to be followed.—B. G. B.

Classification of Vapour Pressure Curves and an Account of some Uses. R. Hörbe and O. Knacke. (*Zeit. Erz. Met.*, 1955, **8**, Dec., 556–561). Curves are given for about 140 metals, oxides, sulphides and halides, including those of iron and of ferrous and ferric-chlorides.

Investigation of the Diffusion of Iron in Iron–Nickel Alloys. M. B. Heiman and A. Ya. Shinyaev. (*Doklady Akad. Nauk SSSR.*, 1955, **102**, 969–972). A series of solid solutions of nickel and Armco iron were annealed and the diffusion coefficient determined from a coating of electrodeposited iron containing ^{59}Fe.

Possible Role of Diffusion in the Creep of Alpha and Gamma Iron. O. D. Sherby and J. L. Lytton. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 2, 928–930). The authors examine the possibility of relating the creep rate of a metal at a given temperature and stress to diffusivity data, referring particularly to α-iron and γ-iron. They find some support for Mott's theory that the rate-controlling process in high-temperature creep is that of dislocation climb.—G. F.

Investigation of Diffusion through the Body of a Metal and along Grain Boundaries by Autoradiography. S. Z. Bokshtein, S. T. Kishkin, L. M. Moroz, and T. I. Gudkova. (*Doklady Akad. Nauk, SSSR.*, 1955, **102**, 73–75). Diffusion of tin into iron and nickel was studied simultaneously. Diffusion into iron was uniform, but in nickel it followed the grain boundaries.

Studies on the Wetting Effect and the Surface Tension of Solids. The Change in Scratch Hardness of Carbon Steel and Aluminium due to Wetting by some Cutting Liquids. M. Sato and K. Miyazawa. (*Proc. Japan Acad.*, 1956, **32**, Apr., 267–269). [In English]. The scratch hardnesses of carbon steel and, to a greater extent, aluminium are lowered by Cimcool and Dash cutting oils, the effect increasing as the surface tension of mixtures of the oil with water increases. This is the opposite of the effect of lubricating oils.—K. E. J.

Oxygen in Liquid Iron–Nickel Alloys. H. A. Wriedt and J. Chipman. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Sept., 1195–1199). The authors have studied the equilibrium in the reaction of hydrogen gas with oxygen in liquid nickel, iron, and their alloys between 1500° C and 1700° C, and have found the equilibrium constant to be greater in nickel than in iron by a factor of 45 at 1594° C. In the alloys, the activity coefficient of oxygen is slightly increased by the presence of nickel.—G. F.

On Oxygen Solubility in Liquid Nickel and Nickel–Iron Alloys. A. M. Samarin and V. P. Fedotov. (*Izvest. Akad. Nauk SSSR, Otdelenie Tekn. Nauk*, 1956, (6), 119–125). [In Russian]. An investigation of oxygen solubility in molten nickel and nickel-iron alloys was carried out. The data obtained were in satisfactory agreement with data obtained by other authors.—V. G.

Hydrogen in Steel. M. Šícha. (*Hutník*, 1955, **5**, (10), 302–305). [In Czech]. The factors determining the hydrogen content are discussed and proposals made for keeping it at a minimum are reviewed with special reference to recent Soviet publications.—P. F.

Formation of Blisters in Iron. F. De Kazinczy. (*Jernkontorets Ann.*, 1956, **140**, (5), 347–359). [In English]. The hydrogen pressure required to initiate cracks associated with blistering was determined for decarburized Armco iron before and after reduction of oxygen content. The samples were annealed at different temperatures after cold rolling. Transcrystalline cracks were observed to start at slag inclusions. The critical hydrogen pressure for their formation depends on the cold strength of the matrix and on flaws introduced during cold working. For crack initiation at grain boundaries it was lower in oxygen-rich iron, due possibly to intergranular brittleness, whilst some cracks were entirely intergranular with no blisters visible at the surface.—G. G. K.

The Physical Metallurgy of Low-carbon, Low-alloy Steels containing Boron. K. J. Irvine, F. B. Pickering, W. C. Heselwood, and M. Atkins. (*J. Iron Steel Inst.*, 1957, **186**, May, 54–67). [This issue.]

Boron in Iron and Steel. G. M. Leak. (*Metal Treatment*, 1956, **23**, Jan., 21–28). A brief review of the effects of boron in iron and steel, with particular reference to its influence on hardenability, is presented. Some of the promising theories which try to explain the mechanism of the hardening effect are discussed. It is shown that there is need for data on diffusion rate and solubility, the type of solution which boron forms and the location of boron atoms in solution. These aspects are also discussed, and some internal friction measurements are described.—A. H. M.

Effect of Small Chromium and Nickel Contents on the Properties of Steel Products. A. Schepers and R. Krauss. (*Stahl u. Eisen*, 1956, **76**, July 12, 896–903). The occurrence and chemical composition of chrome-bearing Conakry iron ores is given. The effect of chromium and nickel contents on the mechanical and other properties of various articles made of Thomas and Siemens-Martin steels was investigated. (Additions of 25, 37·5, and 50% of Conakry ore were made to the blast furnace burden in producing the iron.) The results of the investigation are given and discussed.

Molybdenum in Low and Medium Alloy Steel Castings. W. J. Jackson. (*Alloy Metals Review*, 1956, **8**, March, 2–8). British and U.S. Standards, high and low temperature mechanical properties and creep tests are tabulated and discussed. Uses are considered, and compositions and properties given of suitable materials.

Tool and Die Steels for Hot Working Operations. E. Johnson. (*Alloy Metals Review*, 1956, **8**, June, 2–10). Compositions and heat treatment, tensile, hardness, and impact properties are tabulated and discussed. Carbides are also referred to.

Study on High-strength Low-alloy Steel. T. Mishima, Y. Mishima, E. Asano, and K. Aoki. (*Tokyo Univ. Fac. Eng. Annual Report*, 1954, 146–155). A manganese steel known as Hi-Stren of the Toto Seiko Co. was subjected to mechanical tests as rolled and as annealed. The material is sorbitic. Heat-treatment was investigated and transformation temperature, quenched and quenched-and-tempered hardnesses and tensile properties were measured.

Stainless Steels Including Other Ferrous Alloys. W. A. Luce. (*Ind. Eng. Chem.*, 1956, **48**, Sept., Part II, 1775–1787). A Review. Corrosion, Mechanical properties, High Temperatures, Welding, and other aspects are covered. (192 references).

The Properties of 2¼% Cr–1% Mo Steel. M. G. Gemmill and J. D. Murray. (*Iron Steel*, 1956, **29**, Apr., 150–152; May, 173–177). The authors discuss the tensile, creep, and

rupture properties of 2¼% Cr–1% Mo steel, an established high-temperature steam plant material. The probable effects of heat treatment, welding and the presence of stress-raisers on the performance of the material under service conditions are considered.—G. F.

Mechanical Properties of a Normalized and Drawn Cast Boron Steel. S. L. Gertsman, D. K. Faurschou, and R. K. Buhr. (*Foundry*, 1956, **84,** Aug., 92–99). Data are given on the mechanical properties of a cast low-carbon, molybdenum–boron steel in the normalized and tempered condition. The variables studied include section size, and rare earth additions. The results show that the yield stress-ultimate tensile strength ratio was equal to that of quenched and tempered conventional steels at the same tensile level, and that increasing the section size from 1 to 5 in. caused a moderate lowering of tensile strength. Rare earth additions had no effect on tensile strength, but raised the over-all average impact strength and lowered the fibrous-brittle transition range. The impact values for the boron steel showed a large scatter but were unaffected by section size.—B. C. W.

How to Use the New Low Nickel, Austenitic Stainless Steel. (*Steel Processing*, 1956, **42,** July, 404–406). The article illustrates the success of the new AISI low-Ni high-Mn steels. The chemical composition and physical and mechanical properties of 201 and 202 alloys are given. Corrosion resistance and fabrication and welding characteristics are discussed and compared with their 18–8 counterparts.—A. H. M.

Low-expansion Co–Fe–Cr Alloys: Consistency at Normal Temperatures. L. G. Davis. (*Iron Steel*, 1956, **29,** Aug., 396–397). Studies conducted at the National Bureau of Standards in Washington shows that certain Co–Fe–Cr alloys containing about 36·6% Fe, 8·9% Cr, and balance Co, have thermal expansions which are very low (less than 1×10^{-6} per ° C) and nearly constant in the range 20–60° C. However, the expansion is very sensitive to small variations in composition, and some of the alloys undergo a transformation on cooling at low temperatures.—G. F.

Studies on the Boron-treated Spring Steel T50 B60. II. Y. Yasuda and K. Suzuki. (*Tetsu to Hagane*, 1956, **42,** May, 416–420). [In Japanese]. TS 50 B60 steel was compared with AISI 5160, SUP 6 and other steels in respect of quenching and tempering, notched tensile strength, low-temperature impact properties, resistance to hot deformation, and fatigue behaviour, and was found to have good properties for spring steel.—K. E. J.

Influence on Vanadium, Molybdenum, Tungsten and Silicon on the Properties of the Hot-Working Tool Steel Containing 0·35% Carbon and 5% Chromium. N. Yamanaka and K. Kusaka. (*Tetsu to Hagane*, 1956, **42,** Apr., 327–333). [In Japanese]. Results are given for the effects of the elements stated on the critical point, S-curves for the transformation of austenite, as-quenched and as-tempered hardness, retained austenite, dimensional changes, and mechanical properties of the steel at high temperatures.—K. E. J.

On the Preparation of Monocrystals of Very Pure Iron and Certain of its Properties. J. Talbot. (*Publ. Inst. Rech. Sid.*, 1956, Series A, (137), Apr., pp. 50). The preparation and analysis of pure electrolytic iron is described and its annealing in hydrogen. Impurities were determined by radioactivation and by microanalytical methods (for C, O, N and S) which are outlived. Large single crystals were then grown, complicated by polygonization which was also studied, and X-ray examination, cathodic hydrogen saturation studies and tensile tests carried out on single crystals and polygonized specimens.

The Mechanical Properties of Quenched and Tempered Medium-Carbon Alloy Steels. J. W. Lodge and G. K. Manning. (*Amer. Iron Steel Inst., Contrib. Met. Steel.* No. 49, 1956, Mar. pp. 65). The authors have investigated the effects of incomplete hardening or slack quenching on the mechanical properties of six compositions of medium-carbon alloy steels, the properties being determined by conventional tensile, compression and torsion tests on hot-rolled bar and by tensile and bearing tests on hot rolled sheet. Some of the samples were fully quench-hardened, and others 50% transformed by controlled cooling and tempered to various hardness levels. The results are tabulated and graphed.—G. F.

The British Iron and Steel Research Association. (*Metallurgia*, 1956, **54,** Oct., 166–172, 182). A number of items of research work now being carried out by the Iron Making, Steelmaking, Mechanical Working, Metallurgy and Plant Engineering Divisions and Chemistry. Physics and Operational Research Departments are briefly considered.—B. G. B.

METALLOGRAPHY

Metallography as an Aid to Electrodeposition. A. Comley. (*Electroplating Metal Finishing*, 1956, **9,** Nov., 355–361). Metallographic technique suitable for electroplated specimens is described with examples of its application to plating problems.

A Recording Camera for Metallography with Polarized Light. B. R. Butcher and I. A. McDonald. (*J. Inst. Metals*, 1956, **84,** July, 453–454). The disadvantages are discussed of commercially available recording cameras when used for metallography, and it is shown how these can be overcome by adapting an R.A.F. type F24 camera for the work. Controls for the camera are described.—B. G. B.

Replica Technique for Systematic Examination of a Surface under the Electron Microscope. F. Nesh. (*J. Appl. Phys.*, 1956, **27,** Oct., 1252–1253). This method, which is applicable to a surface forming part of an inner ring, is based on the breaking away of a polystyrene film from the metal. The plastic is applied in 1, 1, 3 and 6% solutions and allowed to air dry for 24–72 hours.—E. E. W.

Equipment for the Examination of the Texture of Coarsely Crystalline Thin Metal Strip. G. Sommerkorn. (*Techn. Mitt. Krupp*, 1956, **14,** Apr., 16–17). Apparatus is described for examining the texture of thin metal strip, its use being illustrated by a photograph of an iron-silicon sample.—L. D. H.

Investigation of Pure Iron and Soft Steel with the Electron Microscope. F. van Wijk and J. A. B. van Dijck. (*Acta Met.*, 1956, **4,** Nov., 657–659). Electron microscope studies of 99·99% Fe and steel containing 0·05% C were made using carbon replicas of electropolished surfaces. The sub-structure was revealed and the effect of slip was also investigated.

A New Method of Micrographic Examination of the Structure of Sintered Carbonyl Iron. G. Cizeron and J. Slobodanka-Mititch. (*Compt. Rend.*, 1956, **242,** Feb., 27, 1177–1180). A method of epitactic oxide formation is described and α- and γ-phases shown.

Metallographic Chemical Polishing of Steel: Variation in Solution Composition with Carbon Content. A. R. Hallett, I. Posener and L. W. Graham. (*Research*, 1956, **9,** Nov., S39). Optimum compositions of oxalic acid–hydrogen peroxide mixtures for steels of different carbon contents are shown.

New Electrolytic Reactives for Stainless Steels Containing the Sigma Phase. J. Bertetti. (*Met. Ital.*, 1956, **48,** July, 324–326). [In Italian]. Some new electrolytic reagents are tested to differentiate the σ phase in stainless, ferritic and austenitic steels. With stainless and ferritic steels fairly good results were achieved by using solutions of ammonium persulphate, sodium perborate, sodium periodate and sodium bromate. With the austenitic sheets good results could only be obtained by using ammonium persulphate and sodium bromate.

Selective Oxidation of Alloying Elements in Steel in DX Gas. D. I. Cameron. (*J. Australian Inst. Met.*, 1956, **1,** Oct., 171–172). DX gas (H_2 14%, CO 8·5%, CO_2 4·5%) and CO–CO_2 and H_2–H_2O mixtures with N_2 were used at 650°, 750° and 880° C. Nine steels and enamel iron (analyses given) were treated and findings are summarized. A full account will appear later.

Cast Structure of High-speed Steel. E. Ineson and G. Hoyle. (*Metal Treatment*, 1956, **23,** July, 257–262). The causes and effects of the formation of coarse cellular carbides in high-speed steels are reviewed. Attention is drawn to possible methods of investigating the problem and eliminating it.

A New Electrolytic Cell for Isolating Carbides and Non-Metallic Inclusions in Steel. N. Backström, S. Heiskanen and U. Ilme. (*Eng. Digest*, 1956, **17,** Dec., 523, 530; from *Jernkontorets Ann.*, 1956, **140,** (10), 812–815).

Development of Procedures for the Identification of Minor Phases in Heat-resistant Alloys by Electron Diffraction. L. O. Brockway and W. C. Bigelow. (*PB* 115971, 1953; *TIDU List* 572, pp. 30). Methods of polishing, etching and rinsing are described. Fe–Cr–Ni–Mo and Fe–Cr–Co–Ni and two other alloys are studied.

A Particular Case of Exogenous Inclusions in Steel. R. Zoja. (*Met. Ital.*, 1956, **48,** July, 321–323). [In Italian]. After a brief review of the subject, the author shows, by means of an example of macroscopic inclusions, the presence of microscopic inclusions in steel. The author has been able to trace their presence from a particular refractory cement.—M. D. J. B.

Ageing of 16-25-6 Steel and Identification of Precipitates.

A. Ferro, S. Gallo and C. P. Galotto. (*Met. Ital.*, 1956, **48**, Aug., 361–368). [In Italian]. The authors study the precipitation of nitrides and carbides in a stainless steel containing nitrogen. (Composition Cr 16%, Ni 25%, Mo 6%, C 0·08%, N 0·15%). The precipitates, extracted by an electrolytic process and observed by X-ray diffraction correspond in most cases to a hexagonal compound probably Cr_2N, with $a = 4{\cdot}79$Å and $c = 4{\cdot}47$ Å, or to a complex carbide of Fe, Mo, Cr with cubic structure with $a = 10{\cdot}96$ Å and isomorphous with some of the high speed steels. These results are confirmed by observations made under the electron microscope. Carbide and nitride precipitation kinetics are studied by the dilatometer process and by accurate measurement of the matrix lattice. Activation energy is found to be about 80,000 cal/gr at. Precipitation is preceded by an induction stage. When the material is cold worked, the precipitation occurs mainly along the slip planes and the induction period practically vanishes. (21 references).—M. D. J. B.

Crystallography and the Phase Rule. A. R. Ubbelohde. (*Brit. J. Appl. Phys.*, 1956, **7**, Sept., 313–321). Pre-transition and pre-melting phenomena, continuous thermodynamic transitions, and hysteresis about a thermodynamic transition are discussed with special reference to single-crystal studies.

Study of the Appearance and Stability of the Polygonal State in Crystals of Pure Iron and Aluminium. Ch. de Beaulieu. (*Publ. Inst. Rech. Sid.*, 1956, Series A, (132), Apr., pp. 42). Properties and microstructures of very pure metals are described and illustrated. Preparation of monocrystals and their polygonization are described and the mechanism discussed. Effects of impurities are studied and the stability of the condition considered. Most of the work described is on Al.

Precipitate Kinetics and Structure in a Cu − 2·4% Fe Alloy. J. M. Denney. (*Acta Met.*, 1956, **4**, Nov., 586–592). The kinetics of the precipitation of iron at 550°–800° C in a Cu −2·4% Fe alloy were studied from measurements of saturation magnetization and were found to proceed by a diffusion-controlled process. The paramagnetic iron precipitate transformed by an unknown martensitic mechanism to a ferromagnetic structure upon plastic deformation but could not be transformed by heat treatment below 800° C. The crystallography of the transformation was studied.—A. D. H.

A Next-Neighbours Theory of Maximum Undercooling. A. S. Skapski. (*Acta Met.*, 1956, **4**, Nov., 583–585). By considering the interactions between next-neighbour atoms and molecules it is shown that for F.C.C. and H.C.P. lattices that the temperature of maximum undercooling = 0·81 × absolute melting point. For other metals and water the theory agrees fairly well with previously determined values for undercooling.

The Direct Observation of Metal Surfaces by Means of Electrons Liberated by Ions. G. Mollenstedt and M. Keller. (*Rev. Univ. Min.*, 1956, 9th series, **12**, Oct., 415–420). An ionic beam directed at an angle of $< 30°$ on to a metallic surface liberates electrons which by using an immersion-type electrostatic or magnetic objective produces a clear image of the surface. The resolving power achieved is 500Å with a theoretical limitation of 100Å. The images possess good contrast if the specimen is maintained at 150° C. The advantages of the technique are illustrated by a number of micrographs obtained from steels and cast irons.—B. G. B.

Evaporated Carbon Film for Use in Extraction Replica Technique. A. Fukami. (*Rev. Univ. Min.*, 1956, 9th series, **12**, Oct., 476–481). A review is presented of experiences gained with this technique for use on precipitates in metals. Details are given of processes used by the author and their application to different steels.—B. G. B.

Some New Problems Arising from Electron Microscope Studies of the Plastic Deformation of Metals. A. F. Brown. (*Rev. Univ. Min.*, 1956, 9th series, **12**, Oct., 501–507). In the development of studies on plastic deformation of metals it is necessary to know whether the phenomena observed on the surface represent truly the interior processes of deformation. Methods of microscopic examination under the surface are discussed.—B. G. B.

Growth of Graphite in Cast Iron. C. E. Birchenall and H. W. Mead. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 2, 1004–1008). The authors calculate the rates of growth of graphite nodules in cast iron from a model of a graphite sphere surrounded by an austenite shell through which carbon and iron are diffusing. The rates calculated from carbon diffusion agree fairly well with measured values, but those based on iron self-diffusion are much too low. Because iron must be removed for the process to continue, a plastic deformation mechanism is suggested.—G. F.

Investigation of the Crystallization of Steel with the Aid of Radioactive Indicators. L. M. Efimov, M. T. Bul'skiĭ, V. I. Yakushin, A. G. Alimov, and A. M. Skrebtsov. (*Stal'*, 1955, (12), 1090–1098). [In Russian]. Radiography of sections has been successfully applied to the study of steel crystallization in ingots. The method indicated reliably the crystallization boundary for the portion of the ingot above the point at which the tracer was introduced. It was found that at each level in the ingot crust, growth in the initial period followed the square-root law " approximately ". The solidification coefficient K depended on the position of the section, falling regularly towards the top where its values were about $1{\cdot}5$ cm/min$^{0{\cdot}5}$. The relation between rate of crystallization and height was confirmed by the temperature distribution in the mould wall at various times after filling.—S. K.

Primary Recrystallization Textures in Cold Rolled Si–Fe Crystals. C. G. Dunn and P. K. Koh. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 2, 1017–1024). The authors have determined the primary recrystallization textures of 11 single crystals of $3\frac{1}{4}$% Si iron cold rolled 70%, and have investigated the large effect of crystal orientation on recrystallization behaviour. Evidence supporting oriented nucleation is given and discussed in terms of probable active slip systems and $< 110 >$ edge dislocations.—G. F.

Effect of the Annealing Temperature on the Formation of the Cube Texture. A. Merlini. (*Trans. Amer. Inst. Min. Met Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 2, 967–972). The author has investigated experimentally the relative merits of two possible explanations of the increase of the fraction of cube texture with increasing annealing temperature of rolled polycrystalline metals. His results are in excellent agreement with the principles of the oriented growth theory, but contradict the oriented nucleation theory.—G. F.

Investigation of Stony Fracture by a Tracer Atom Method. S. S. Nosyreva and A. M. Polyakova. (*Stal'*, 1955, (12), 1120–1123). [In Russian]. Radioactive sulphur ^{35}S was used to study the causes and characteristics of the stone-like fracture in structural steels which is produced by mechanical working at very high temperatures. It was found that in fractured specimens cooled slowly after overheating and then subjected to normal heat treatment, the sulphide inclusions exist either as large accumulations forming a discontinuous network at austenite grain boundaries or as smaller ones inside grains. In samples cooled rapidly after overheating, which do not give the fracture, the sulphides are distributed throughout the grain as small inclusions. Thus solution of sulphides during overheating followed by their separation at the grain boundaries is the probable cause of stone-like fracture.—S. K.

Grain Boundaries. D. McLean. (*Metal Treatment*, 1956, **23**, Jan., 3–10; Feb., 55–61; Mar., 91–96). A comprehensive survey is presented of the present state of knowledge of the various aspects of grain boundaries. First, the author deals with the structure and energies of grain boundaries and discusses the microstructure and shapes of grains for both single-phase metals and two-phase alloys. The equilibrium segregation to grain boundaries is then discussed together with its control by heat treatment. The thermodynamical theory of the formation of such grain boundaries is also put forward. The author finally considers the influence of boundaries on strain hardening. The barrier and the complexity effect are discussed, as well as the strain hardening in different polycrystalline metals. The significance of grain size is referred to.—A. H. M.

Grain Boundaries and the Constituents of Grain Boundaries in Technical Steels. H. K. Görlich. (*Rev. Univ. Min.*, 1956, 9th series, **12**, Oct., 465–470). The author discusses the nature, properties, and importance of boundaries in metallic polycrystals. The classical methods of examination of boundaries and their constituents are described together with a new technique of isolation which conserves the original position of these constituents.—B. G. B.

Structure and Energy of Grain Boundaries. III. F. Erdmann-Jesnitzer. (*Hutnické Listy*, 1956, **11**, (7), 424–432). [In Czech]. The paper represents the final instalment of a detailed review of present-day knowledge relating to grain boundaries, particular reference being made to the author's

recent researches. The technological significance of theory is shown by relating it to weld-embrittlement, cavitation, intercrystalline corrosion, grain coarsening on sintering under pressure, etc.—P. F.

Morphological Study of Rupture Surfaces by the Electron Microscope (Microfractography). J. Plateau, G. Henry, and C. Crussard. (*Rev. Univ. Min.*, 1956, 9th series, **12,** Oct., 543–554). The application of a carbon replica technique on fracture surfaces shows the different types of surfaces. The classical rupture types (ductile, intergranular, cleavage) and fatigue fractures have been examined and 32 photomicrographs of fracture surfaces are reproduced.—B. G. B.

Some Considerations on the Determination of Transformation Points of Steels by the Differential Dilatometer Method. J. Papier. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1956, No. 9, 1831–1835). The Chevenard differential dilatometer technique is briefly described and the interpretation of the results obtained and the determination of the transformation points of steels are explained. Examples of results obtained with Mn–Si steel and a self-hardening steel are given.—B. G. B.

A Contribution to the Study of the Decomposition of Austenite during Step-Quenching. B. Estruch Subirana. (*Inst. Hierro Acero*, 1956, **9,** May, 595–613. Special Number). [In Spanish]. The effect of holding specimens of nine different nickel–chromium steels for various times at chosen subcritical temperatures has been studied in relation to the subsequent transformation process at lower temperatures. When both temperatures lie in the upper, pearlitic, range the reaction times are approximately additive; this is not so if one or both of the subcritical temperatures lie in the bainitic range. An explanation, based on the theory of athermal nucleation of bainite, is offered. (20 references).—P. S.

Transformation of Austenite during Continuous Cooling in Nine Hypoeutectoid Nickel–Chromium Steels. B. Estruch Subirana. (*Inst. Hierro Acero*, 1956, **9,** July, 871–882. Special Number). [In Spanish]. Continuous cooling transformation diagrams have been determined for these steels and the results compared with the isothermal transformation diagrams determined previously. This comparison was made by means of the Scheil hypothesis and it was observed that the formation of ferrite or pearlite occurred as an additive reaction whilst bainite formation was completely unpredictable. It is concluded that further knowledge of the characteristics of isothermal transformation is necessary before continuous cooling behaviour can be predicted.—P. S.

A Study of the Isothermal Transformation of Austenite in 9 Hypoeutectoid Nickel–Chromium Steels. B. Estruch Subirana. (*Inst. Hierro Acero*, 1956, **9,** Apr., 432–462. Special Number). [In Spanish]. The experimental methods used for the determination of the isothermal transformation diagrams of the steels are described, and the diagrams are given. (The analyses range from 0·30 to 0·52% C, 0·94 to 3·29% Ni, and 0·05 to 3·28 Cr). A critical review of the existing theories of the transformation kinetics is given in relation to the author's observations. The experimental data concerning transformation are summarized and an explanation of the mechanism is discussed. (49 references).—P. S.

Effect of Pre-superheating of Steel on the Kinetics of Breakdown of Supercooled Austenite. V. D. Sadovskiĭ. (*Doklady Akad. Nauk SSSR*, 1955, **102,** (3), 515–517). Steels quenched from 650°, one sample being preheated to 1300°, were compared. The second samples showed large grains and the troostite deposited on the boundaries whereas the first samples were almost completely broken down when examined. Magnetometric measurements at 650° are given.

Investigations of the Kinetics of Martensite Formation. H. Beisswenger and E. Schell. (*Arch. Eisenhüttenwesen*, 1956, **27,** June, 413–420). Apparatus for the measurement of rapidly completed transformations is described, and electrical resistance measurements on iron–nickel alloys are reported. The results are discussed and some anomalies are mentioned.

Phenomena in the Heat Treatment of Martensite. W. Jellinghaus. (*Arch. Eisenhüttenwesen*, 1956, **27,** July, 433–448). The heat treatment of martensite was investigated by carrying out measurements of the magnetisation, electrical resistance, coercivity, and volume change in three unalloyed steels containing 0·2, 0·4, and 0·8% carbon. The experiments were mainly conducted under constant temperature conditions sustained over lengthy periods. The results are given and discussed in some detail.

On Banding in Steel. C. F. Jatczak, D. J. Girardi, and E. S. Rowland. (*Trans. Amer. Soc. Metals*, 1956, **48,** 279–305). The banding characteristics of 1340, 2340, 4140, 4340, and 5140 type steels were investigated by studying the transformation behaviour to ferrite and pearlite and to martensite on direct quenching and indicated that chemical heterogeneity was the primary cause of the phenomenon.—E. E. W.

The Tempering of Iron-Carbon Martensite Crystals. F. E. Werner, B. L. Averbach and M. Cohen. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 43). The first stage tempering reaction is a continuous formation of ε-carbide and martensite with about 0·30% C; the third stage reaction is continuous with the C gradually draining from the matrix solution. Broadening due to particle size rapidly disappears during the third stage, while the strain persists to high temperatures. The character of the strain relief appears similar to that of recovery in cold-worked metals.—E. E. W.

X-Ray Diffraction Studies of the Bainite Transformation in Four Alloy Steels. L. S. Birks and R. T. Seal. (*PB* 111771, 1955, pp. 11; *TIDU List* 579). Isothermal and continuous cooling studies show that with a 3 sec austenitizing time, the transformation is more rapid the lower the austenitizing temperature for a steel without boron but possibly reversed if B is present. A rapid graphical solution of the Scheil approximation is given and predicted and observed transformations were found to be in good agreement for 2 steels.

The Influence of Bainite on Mechanical Properties. R. F. Hehemann, V. Luhanov and A. R. Troiano. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 16). At constant strength level, high temperature bainite lowers yield strength, reduction in area and impact energy of 4340 steel heat-treated to high strengths. As the temperature of formation of the bainite is lowered, damage to mechanical behaviour becomes less severe.—E. E. W.

The Chemical Nature of the Sigma Phase in the Fe–Cr Systems. A. T. Grigor'ev, N. M. Gruzdeva and I. A. Bondar. (*Investiya Sektora Fiziko-Khimicheskogo Analiza*, 1954, **21,** 132–143). Preparation and annealing of Fe–Cr and Fe–Cr–2% Ni alloys are described and the sigma phase discussed in the light of hardness measurements and thermal analysis. Micrographs and magnetic properties are also examined.

Phase Relationships and Mechanical Properties of Some Iron–Chromium–Carbon–Nitrogen Alloys. G. F. Tisinai and C. H. Samans. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 30). Alloys cooled from 2200° F produced austenite to varying degrees and the carbides and nitrides were largely dissolved. At temperatures below 2200° F the solubility of C and N decreased causing precipitation of carbides and nitrides and forming some ferrite which varied inversely with temperature and at 1500° F comprised the complete matrix in all alloys. On rapid cooling from 1500–2000° F the austenite tended to transform to martensite. Tensile tests at room and elevated temperatures showed that the austenised alloys had much higher strengths than the austenitic Fe–Cr–Ni steels but still had good ductilities. Impact tests proved that there were ductile-brittle transition temperatures.—E. E. W.

Metallography of Titanium-Stabilised 18-8 Stainless Steel. T. V. Simpkinson. (*Trans. Amer. Soc. Metals*, 1956, **49,** Preprint 29). Results of tests on AISI 321 are illustrated by photomicrographs and discussed. Time and temperature conditions for the production of ferrite, "new" austenite, sigma, chromium carbide and titanium carbide are exemplified.—E. E. W.

Zirconium. Its Production and Properties. (*U.S. Bur. Mines, Bull.* 561, 1956, pp. 180). The Zr–Fe systems and methods of determination of iron in zirconium alloys are included.

CORROSION

Fundamentals of Liquid Metal Corrosion. W. D. Manly. (*Corrosion*, 1956, **12,** July, 336–342). The "corrosion" of solid metals by liquid metals is mostly simple dissolution. This process is often complicated and assisted by alloying between components, intergranular penetration, impurity reactions and mass transfer resulting from temperature gradients or concentration gradients between dissimilar metals. The various types of attack experienced are illustrated and discussed. The important roles of temperature and temperature differences are explained, but the rate controlling factors have, in most cases, still to be discovered.—J. F. S.

Effects of Metal Cations on the Corrosion of Iron in Acids. H. C. Gatos. (*Corrosion*, 1956, **12**, July, 322t–330t). The corrosion of mild steel in 1N sulphuric, hydrohloric and acetic acids was found to be increased by the presence in solution of Fe^{+++}, Ce^{++++}, Cu^{++} and $Cr_2O_7^{--}$ ions. The corrosion rate increased linearly with concentration in [the range chiefly studied (up to 0·1 g ions per litre) but polarization of local anodes probably accounted for the smaller increase for $Cr_2O_7^{--}$ at higher concentrations. As^{++} and Sn^{++} inhibited the corrosion effectively. Fe^{++}, Cr^{+++} and Mn^{++} had no effect on the corrosion rate. Ni^{++} unexpectedly accelerated corrosion by being reduced by the iron. Mechanisms for the various reactions are elucidated by reference to the electrode potential measured in addition to the weight loss.—J. F. S.

Corrosion Research at the Avesta Steelworks Laboratories (Sweden). T. G. Owe Berg. (*Acciaio Inossid.*, 1956, **23**, May, Aug., 47–87). [In Italian]. The author describes researches carried out at Avesta on the physico-chemical aspects of concentrated solutions of acids, on the physical behaviour of metals and on the kinetics of the reactions of these metals when in contact with the acids.—M. D. J. B.

Symposium on High-Purity Water Corrosion. (*Amer. Soc. Test. Mat. Special Publ., No.* 179, 1956). **Influence of Water Composition or Corrosion in High Temperature, High-Purity Water.** D. M. Wroughton, J. M. Seamon and P. E. Brown, (19–26). Carbon steels in nuclear reactors are considered. **Effect of Material Composition in High-Temperature Water Corrosion.** A. H. Roebuck, (27–36). Steels and non-ferrous alloys are reviewed. **Special Corrosion Study of Carbon and Low Alloy Steels.** R. U. Blaser and J. J. Owens. (37–54). Apparatus simulating boiler conditions is described.

The Influence of Some Factors, Particularly in Basic Bessemer Steel, on Tinplate Quality. J. Teindl. (*Pišek's Sborník*, 1956, 263–270). [In Czech]. Corrosion of tin cans is investigated with regard to type of steel strip used. Basic Bessemer and O.H. are compared and the former found satisfactory. It is recommended that a small amount of Si (0·08–0·1%) be added to improve the surface of the sheet and decrease the thickness of tin.

A Study of Nonprotective Rust Formation on Auto-Body Sheet Steel. J. C. Holzwarth, R. F. Thomson and A. L. Boegehold. (*Trans. Soc. Auto. Eng.*, 1956, **64**, 221–228). Accelerated corrosion tests are described and illustrated in colour. Effects of alloying elements are briefly considered.

Corrosion in Caustic of Nickel-Iron Welds obtained in Fabrication of Nickel-Clad Vessels. P. J. Gegner. (*Corrosion*, 1956, **12**, June, 261t–262t). Up to 25% iron can be tolerated in the welds of nickel-clad vessels so far as corrosion resistance to 73% caustic soda at 250° F is concerned.—J. F. S.

Corrosion of Nickel-Chromium-Iron Alloys by Welding Slags. G. R. Pease. (*Welding J.*, 1956, **35**, Sept., 469s–470s). The nature of the attack is described, under both oxidizing and reducing conditions and the temperature limits defined within which damage can be expected.—U. E.

Galvanic Corrosion in Water Meters. H. F. Barrett. (*J. Amer. Water Works Assoc.*, 1956, **48**, Oct., 1235—1242). Eleven galvanic couples occurring are mentioned and their effects briefly discussed.

Tuberculation of Tar-coated Cast Iron in Great Lakes Water. T. E. Larson, R. V. Skold and E. Savinelli. (*J. Amer. Water Works Assoc.*, 1956, **48**, Oct., 1274–1278). A preliminary report.

High Temperature Corrosion of Metals by Vanadium Dioxide. Researches on Metal–Vanadium–Oxygen Systems. V. Cirilli, A. Burdese and C. Brisi. (*Met. Ital.*, 1956, **48**, July, 309–316; Aug., 349–356). [In Italian]. The authors study the corrosion caused by vanadium dioxide at high temperatures and establish equilibrium diagrams, up to now unknown, of Fe, Cr, Ni and Co oxides. In Cr_2O_3–V_2O_5 and Fe_2O_3–V_2O_5 systems only one compound appears namely orthovanadate. But in CoO–V_2O_5 and NiO–V_2O_5 systems the ortho-, meta- and pyrovanadates are all found. Finally the authors study the coexistence relations at 700° C of the phases appearing in the ternary systems formed by the four metals. (26 references).

Contamination and Corrosion in Rail Tank Cars. S. J. Oechsle, jun. and K. G. Le Fevre. (*Corrosion Techn.*, 1956, **3**, Dec., 389–392). The authors enumerate and briefly discuss the factors controlling corrosion in rail tank cars and show how linings (thin, thick, combination and sheet linings) may be used to prevent such corrosion. The compositions, pretreatments and methods of applying these linings are given and the surface preparation of areas to be lined is described. It is concluded that a protective lining is the best method of overcoming corrosion.—L. E. W.

Pitting Corrosion of Reserve Fleet Ships. E. F. Corcoran and J. S. Kittredge. (*Corrosion Prevention and Control*, 1956, **3**, Dec., 45–48). Inactivated ships of the U.S. Pacific Reserve Fleet moored in San Diego Bay are being heavily damaged by intensive pitting corrosion. Chlorinity determinations, drift spar measurements and oxygen and carbon analyses in conjunction with phytoplankton counts, microchemical analysis and a study of microbiological cultures point to electrochemical corrosion in oxygen concentration cells created by rapid bacterial oxidation of organic matter absorbed by ships' plating. Wavelet flushing at the water line with water supersaturated with oxygen provides the cathodic area. Maximum corrosion rates are then maintained by sulphate-reducing bacteria that provide sulphur for anodic depolarization. (13 references).—L. E. W.

The Corrosion of Cargo Ships and its Prevention. H. J. Adams and J. C. Hudson. (*Trans. Inst. Marine Eng.*, 1956, **68**, Nov., 429–448; discussion 448–459). A comprehensive review of corrosion in various locations and of painting systems is given.

The Performace of a Cathodic Protection System on the U.S. Ship "Sumner," Dec. 1953–Jan. 1955. J. Waldronnl and H. M. Peterson. (*PB* 118289, 1955, pp. 6; *TIDU List* 584). Magnesium anodes and an impressed-current anode system were used. The installation, potential current daily readings and hull inspection are described. The protection of the hull was estimated at 95% and of the rudders, shafts and struts at 85%.

The Performance of a Cathodic Protection System on the LST 532. L. J. Waldron and H. M. Peterson. (*PB* 118288, 1955, pp. 5; *TIDU List* 584). A Mg-anode system is described cast round galvanized steel grids. Daily potential measurement are shown and the system was completely successful from April 1953–April 1955.

The Inhibitor I.F.P. 550. (*Indust. Finishing*, 1956, **9**, Sept., 109–113). The development and properties of an inhibitor which is a twice esterified (benzyl-xanthate) glycol are described. The influence of the chemical structure on inhibitory properties is fully discussed and the effect of the inhibitor in preventing the attack of 18% HCl on steel at 20°, 60° C and 100° C is given.—A. D. H.

Protecting Ammonium Sulphate Crystallizers by Means of Inhibitors. G. Roberti, F. Gianni and G. Bombara. (*Met. Ital.*, 1956, **48**, June, 281–286). [In Italian]. The authors describe the corrosion which takes place in stainless steel salt crystallizers when producing ammonium sulphate. In their attempts to find a means of protection they studied the influence of the addition of some fifteen different materials to the solutions. The authors establish that copper sulphate, potassium dichromate, hydrogen peroxide and sodium arsenite possess a very good inhibiting action. (9 references).—M. D. J. B.

Two-Step Process Retards White Rust. L. J. Brown. (*Iron Age*, 1956, **178**, Nov. 8, 106–109). A two-dip treatment for galvanized steel is described. Testing is described. A water-repellent film is formed and dipped in a chromic-type bath.

Laboratory Method for Corrosion Inhibitor Evaluation. G. A. Marsh and E. Schaschl. (*Corrosion*, 1956, **12**, Nov., 534–538). A screening test for inhibitors is described. A thin steel test probe is inserted in a rotating bottle containing oil-brine-gas or similar multiphase corrodants selected to resemble actual conditions. The A.C. conductance of the temperature-compensated test strip is measured either intermittently or continuously. An application of the method in testing rust inhibitors for gasoline is discussed.—J. F. S.

Temporary Corrosion Preventives. (*Corrosion Prevention and Control*, 1956, **3**, Oct., 31–32, 46). The properties and applications are described of soft- and hard-film solvent-deposited preventives, soft-film hot-dipping, grease and slushing preventives, strippable hot-dip coatings, vapour phase-inhibitors and cocooning.—L. E. W.

ANALYSIS

French Standard Samples in the Iron and Steel Industry. M. E. Jaudon. (*Chim. Analy.* 1956, **38**, June, 195–200). A general account is given of the IRSID standard steels, how they came to be issued, their preparation and uses.

Identification of Colours in Spot Analysis. H. Malissa.

(*Arch. Eisenhüttenwesen*, 1956, **27**, Mar., 177–178). One of the difficulties in the application of spot analytical techniques lies in the identification of colours and in the determination of colour intensity. A method is described by which colour and colour intensity can be characterized, and the determination of nickel is given as an example.

Determination of Phosphorus in the Field and in Iron Ore Stockpiles. U. Fernlund, S. Zechner, and S. O. Eriksson. (*Jernkontorets Ann.*, 1956, **140**, (5), 389–391). [In Swedish]. Details are given of a qualitative and a semi-quantitative method for determining phosphorus, by precipitating as quinoline phosphomolybdate, from which it is easy to differentiate between ores with various phosphorus contents. If the indicator solution is sprayed directly on to the surface of the ore, apatite becomes yellow in colour.—G. G. K.

The Colorimetric Determination of Phosphorus in Steel and Copper-base Alloys. W. T. Elwell and H. N. Wilson. (*Analyst*, 1956, **81**, Mar., 136–144).

Coulometric Volumetric Method for Determination of Silicon and Phosphorus in Iron and Steel. W. Fuchs and O. Veiser. (*Arch. Eisenhüttenwesen*, 1956, **27**, July, 429–432). After a short review of the literature the authors outline the coulometric method. The apparatus is described, and details of the automatic titration are given. Determinations are said to take 13–15 minutes.

Simultaneous Determination of Small Carbon and Nitrogen Contents in α-Iron by Measurement of Damping. W. Wepner. (*Arch. Eisenhüttenwesen*, 1956, **27**, July, 449–452). In conjunction with earlier work on the effect of carbon dissolved in α-iron on damping, the relationship between damping and nitrogen content in α-iron was investigated. On the assumption that there is no interaction between carbon and nitrogen, a formula is derived and made the basis of a method of measuring damping and thereby obtaining simultanously the carbon and nitrogen contents. The magnitude of the possible error is estimated.

Contribution to the Study of the Determination of Nitrogen in Cast Iron. J. Marot. (*Rev. Mét.*, 1956, **53**, Sept., 689–700). The development of a method of nitrogen analysis by vacuum fusion in the presence of sodium peroxide, suitable for us with cast irons, is described. Results obtained are used to explain discrepancies obtained by the classical Kjeldahl technique which does not determine all the nitrogen present. Heat treatment of the specimens at 950° C for 3 hours has been found to enable the Kjeldahl technique to give results comparable with the new vacuum oxidation method.—B. G. B.

Nitrogen Analysis in Steels by the Wet Process. J. Calmettes and J. Drain. (*Rev. Mét.*, 1956, **53**, Sept., 682–688). The absorptiometric analysis of ammonia, distilled from steel specimens, using the formation of a blue indophenol by reaction of ammonium ions with a solution of sodium phenate–hypochlorite mixture is described in detail. The method has been found sensitive, selective and rapid. A comparative study of acid-soluble and total nitrogen analyses on carbon steels has shown systematic differences. The titrimetric analysis of the ammonia solution was found to give erratic results.—B. G. B.

Predicting the Solubility of Nitrogen in Molten Steel. F. C. Langenberg. (*Trans. Amer. Inst. Min. Met. Eng.*, 1956, **206**; *J. Met.*, 1956, **8**, Aug., Section 1, 1099–1101). The author presents a method for computing the solubility of nitrogen in molten alloy steels. Examples illustrating the procedure are given, and comparisons are made between predicted and measured nitrogen solubilities.—G. F.

Absorption-Titration Flask for Determination of Sulphur in Steel. M. Roth and S. Lader. (*Anal. Chem.*, 1956, **28**, Aug., 1359).

Photometric Determination of Silicon in Steels. T. R. Andrew and C. H. R. Gentry. (*Analyst*, 1956, **81**, June, 339–348).

Determination of Iron in Titanium Sponge, Alloys, and Ores. H. V. Malmstadt and C. B. Roberts. (*Anal. Chem.*, 1956, **28**, Sept., 1412–1416). A method of automatic derivative spectrophotometric titration with electrolytically generated titanous ion is described.

The Chlorine Method for the Determination of Non-metallic Inclusions in Steel. C. W. Short, R. S. Roberts, and G. Croall. (*J. Iron Steel Inst.*, 1957, **186**, May, 85–89). [This issue.]

A Rapid Photometric Method Determining Small Quantities of Iron in the Presence of Nickel and Chromium. O. Běloh-lávek. (*Chem. Listy*, 1956, **50**, (7), 1195–1197). [In Czech].

The Application of Flame Photometry to the Analysis of Some Elements in Ferrous Alloys. F. Burriel Marti, J. Ramirez Muñoz, and M. del C. Asunción Omarrementeria. (*Inst. Hierro Acero*, 1956, **9**, Apr., 417–424. Special Number). [In Spanish]. An account of the estimation of cobalt, nickel, chromium, and manganese by means of a Beckman DU flame photometer is given. The interference of iron is prevented by a previous separation with organic solvents. An examination of the influence of the presence of these elements upon each other indicated the advisability of separation of the nickel; this was done by a double dimethylglyoxime precipitation. Correction curves were also derived. The results of these estimations are given in comparison with chemical determinations and show a reasonable agreement. (11 references).—P. S.

Direct Photometric Determination of Aluminium in Iron Ores. U. T. Hill. (*Anal. Chem.*, 1956, **28**, Sept., 1419–1424).

Determination of Copper in Steel. L. J. A. Haywood and P. Sutcliffe. (*Analyst*, 1956, **81**, Nov., 651–655). A method is described using bis*cyclo*hexanone oxalyldihydrazone as a colorimetric reagent. It may be used as a spot test and may be applicable to other alloys.

Argentic Oxide as an Oxidimetric Reagent. Determination of Manganese, Cerium and Chromium. J. J. Lingane and D. G. Davis. (*Anal. Chim. Acta.*, 1956, **15**, 201–206). The method can be applied to iron alloys.

Polarography and the Use of 1 : 2-*cyclo*Hexanedione dioxime (Nioxime) in the Amperometric Titration of Nickel and Cobalt. C. Calzolari and C. Furlani. (*Publicazioni delle Facolta di Scienze e di Ingegneria dell' Universita di Trieste*, Series B. 1953, (123), pp. 26).

Quantitaive Inorganic Chromatography. II. The Determination of Vanadium in the Presence of Excess of Dichromate. F. H. Pollard, G. Nickless, and A. J. Banister. (*Analyst*, 1956, **81**, Oct., 577–581).

Rapid Mineral Identification by Electrographic Contact Printing. K. F. G. Hosking. (*Mining Mag.*, 1956, **94**, June, 335–338). Polished sections of ores or embedded grains are identified by a test suitable for the field, the laboratory, and the mill. Arsenopyrite can be detected in this way.

Recommended Methods for the Analysis of Trade Effluents. Methods for the Determination of Iron, Mercury, and Nickel. A.B.C.M.–S.A.C. Joint Committee. (*Analyst*, 1956, **81**, Mar., 176–180).

Possibilities of Direct Reading Spectroscopy For Steels and Cast Irons. A. Camuñas Puig. (*Inst. Hierro Acero*, 1956, **9**, Apr., 405–416. Special Number). [In Spanish]. The principles and operation of the quantometer are described and its uses for the routine analysis of alloys and in research work are discussed. (14 references).—P. S.

An Examination of the Unicam S.P.600 Spectrophotometer. D. R Curry and P H. Scholes. (*Metallurgia*, 1956, **54**, Sept., 145–149). A simplified spectrophotometer for accurate colorimetric measurements within the visible and near infrared regions of the spectrum is briefly described. Certain features of its performance and design have been examined under practical test conditions and some assessment of its potential application in metallurgical analysis is made.

Spectrophotometer Saves Time in Analysis of Grey Iron. (*Foundry*, 1956, **84**, Oct., 172). The use of a spectrophotometer for the rapid analysis of grey cast iron is very briefly described.—B. C. W.

A Universal Method of Spectrochemical Analysis. (*Amer. Soc. Test. Mat. Bull.*, 1956, Sept., 29–32). A progress report of an A.S.T.M. committee to establish an approved spectrochemical method for analysing any miscellaneous sample is presented. Published methods are reviewed and some of the results obtained are discussed. (39 references).—B. G. B.

Spectrophotometric Studies on Organometallic Complexes used in Analytical Chemistry. II. Composition and Dissociation Constant of Ferrous Nitroso R Salt Complex. Y. Oka and M. Miyamoto. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1955, **7**, Oct., 482–486). Dissociation constants of the stable S complex and of nitroso R salt have been evaluated, and the compositions of three of the four types of complexes obtained in the reaction of nitroso R salt with ferrous iron were studied by the molar ratio method.—J. G. W.

Rapid Determination of Metals in Organic Products with Alumina as Spectrographic Aid. P. L. Work and A. L.

Juliard. (*Anal. Chem.*, 1956, **28,** Aug., 1261–1264). Methods are given for Ba, Ca, Fe, Pb, Ni, and Zn.

The Method without use of a Comparison Sample. N. W. H. Addink. (*Lab. N. V. Philips' Gloeilampfabrieken, Separaat*, 2285, pp. 6, from *Mikrochemica Acta*, 1955, (2–3), 703–708). A review of the spectrochemical determination of traces of iron and many other elements in minerals. With a discussion.

Spectroscopic Determination of Phosphorus in Steel. H. Krempl and F. Bertram. (*Arch. Eisenhüttenwesen*, 1956, **27,** May, 303–309). Literature is surveyed for matter relating to the degree of resolution possible in the detection of phosphorus. The sensitization of plates, optimum choice of excitation, surface preparation of specimens, and the general method of experimentation are discussed.

Ion Exchange—Spectrophotometric Determination of Aluminium. A. D. Horton and P. F. Thomason. (*Anal. Chem.*, 1956, **28,** Aug., 1326–1329). The method is applicable to stainless steel and iron alloys as well as to non-ferrous materials.

Theoretical Chemistry of Metals. V. F. Mott. (*Nature*, 1956, **178,** Dec. 1, 1205–1207). Wave functions and the Fermi surface are discussed with notes on the deHaas-van Alphen effect and its interpretation.

Modern Methods of Physico-Chemical Analysis of Iron and Iron Alloys. R. Faivre. (*Annales de l'Est, Faculte'des Lettres de l'Université de Nancy. Mem. No.* 16, 1956. *Rept. of Internat. Colloquium, Nancy*, Oct., 1955, 567–578). **Methods of Chemical Analysis of Iron Ores.** J. Aubry. (579–582). Brief reviews are given of metallographic and other studies of transformations in iron, and of chemical methods.

Recommended Methods for Quantitative Spectrographic Analysis of Low Alloy Steels. (*Met. Ital.*, 1956, **48,** Aug., 281–286). [In Italian]. Details are given of the methodology proposed by the A.I.M. Centre for Spectrographic Research. Specifications are given for the preparation of the sample, the apparatus to be used, the photographic material and method of handling it, the analytical procedure to be followed, the reproducibility and degree of accuracy demanded. Tables and spectra are given indicating the standards to be achieved.

Determination of Oxygen in Steel by the Aluminium Diffusion Method. C. G. Carlsson. (*Jernkontorets Ann.*, 1956, **140,** (8), 551–574). [In English]. Detailed tests were made of Gotta's aluminium diffusion method for determining oxygen in steel and several sources of error were observed which might give rise to serious errors with killed and high-carbon steels. Such error sources have now been eliminated by modifying the method in several important respects making it at least equal to vacuum fusion in all respects except speed and universality. A further analysis was made of the steel samples examined in the international co-operative study of 1934–35 and the results are deemed more reliable than those previously reported. A detailed description is given of the method, together with the results obtained during spectrophotometric determination of aluminium in isolated aluminium oxide.

Spectrochemical Determination of Phosphorus in Steel. E. S. Kudelya and A. S. Dem'yanchuk. (*Doklady Akad. Nauk S.S.S.R.*, 1952, **83,** (37, 397–398).

The Determination of Phosphorus in Low Carbon Ferro-Chrome. B. Bagshawe and A. L. Pill. (*Metallurgia*, 1956, **54,** Nov., 251–254). The inhibitive effect of high chromium concentrations on the precipitation characteristics of ammonium phosphomolybdate are explained and a method based on the British Standard for steels after a preliminary volatilization of chromium as chromyl chloride is recommended for the determination of phosphorus in low carbon ferro-chrome.—B. G. B.

Spectrophotometric Determination of Silicon and Phosphorus in Steels Using Stannous Oxalate as Reductant. C. O. Ingamells. (*Chem.-Analyst*, 1956, **45,** Mar., 10–11).

Colorimetric Determination of Traces of Antimony in Steel. N. Leontovitch. (*Chim. Analy.*, 1956, **38,** Aug., 284–288). The use of an iodoantimonious acid is described in presence of Mn, Cr, W, and V. Mo interferes but can be corrected for. As and Sn are without effect, but the method fails in presence of Se or Te.

Note on the Determination of Traces of Aluminium in Steel. R. Rosotte. (*Chim. Analy.*, 1956, **38,** July, 250–252). A cupferron method with extraction of the iron and nephelometric determination of the Al is described.

Determination of Arsenic in Steel by X-ray Fluorescence. M. B. Cavanagh. (*PB* 111870, 1955, pp. 6; *TIDU List* 579). Germanium is used as collector and internal standard and the sulphide precipitate prepared and examined.

Note on the Potentiometric Determination of Chromium and Vanadium in Steels. J. Silvestre and M. F. Hurth. (*Chim. Analy.*, 1956, **38,** July, 253–257). A review of other methods is included and the effects of tungsten are considered.

Rapid Determination of Silicon in Ferrosilicon. G. Hirschfeld. (*Anais Assoc. Brasil, Quim.*, 1953, **12,** Mar., 25–28). A method taking 2½ hours is given involving removal of silicon as SiF_4. It is reasonably comparable with fusion methods.

The Use of Complexone in the Gravimetric Determination of Titanium in Ferrotitanium. K. Protiva. (*Hutnické Listy*, 1956, **9,** (8), 486–487). [In Czech].—P. F.

Iodimetric Determination of Vanadium and Copper when Present Together. M. R. Verma and V. M. Bhuchar. (*J. Sci. Indust. Res.*, 1956, **15B,** Aug., 437–443). Vanadium may be determined more accurately by the iodimetric method than by the sulphur dioxide method. Tests were made using concentrations of 2·5 to 26 mg V and 6 to 66 mg Cu and also when proportions of V to Cu were 5 : 1 and 1 : 25. Optimum conditions are stated.—E. E. W.

On the Rapid Determination of CaO in Iron-rich Raw Materials. H. Ihara and H. Hoketsu. (*Sumitomo Metals*, 1956, **8,** July, 178–184). [In Japanese]. Normal procedures for determining CaO in iron-containing raw materials, e.g. sinters produced with added lime, suffer from the disadvantage of having to eliminate iron to prevent interference by Fe^{III}. This is removed by a procedure based on conversion of Fe^{III} to a complex salt by acetic acid; CaO contents are determined in 14 min.—K. E. J.

Determination of Iron, Titanium, Phosphorus and Aluminium in Zirconium Ores. P. Santini and F. Diehl. (*Anais Assoc. Brasil, Quim.*, 1952, **11,** 167–171).

The Separation of Iron Valencies by Paper Chromatography. H. M. Stevens. (*Anal. Chim. Acta*, 1956, **15,** Dec., 538–542).

Paper Chromatography of Inorganic Ions. XV. R_F Values in Mixtures of Butanol and HBr. S. Kertes and M. Lederer. (*Anal. Chim. Acta*, 1956, **15,** Dec., 543–547). Ferric and 53 other ions are measured.

Spectrophotometry of the Ferrate (VI) Ion in Aqueous Solution. Z. G. Kaufman and J. M. Schreyer. (*Chem.-Analyst*, 1956, **45,** Mar. 22–23).

Photocolorimetric Methods in Metallurgical Analysis. V. Mayer. (*Hutnik*, 1956, **6,** (7), 209–212). [In Czech]. The development, scope and uses of photocolorimetric methods and instruments are considered.—P. F.

Metallic Transfer in Engineering Operations. Radioactive Methods Used at Cambridge. F. P. Bowden and J. B. P. Williamson. (*Engineering*, 1956, **182,** Nov. 16, 619–621). A review of the results obtained by autoradiography on friction and impact, their significance, and the effects of lubrication etc.

INDUSTRIAL USES AND APPLICATIONS

Symposium on Metallic Materials for Service at Temperatures above 1600° F. (*Amer. Soc. Test. Mat. Special Publ. No.* 174, 1956). **Introduction.** V. N. Krivobok, (1–2). **Some Sheet and Bucket Materials for Jet-engine Applications at 1600° F and Higher.** J. P. Denny, L. P. Jahnke, E. S. Jones, and F. C. Robertshaw, jun., (3–15). **Titanium-carbide-base Cermets for High-temperature Service.** K. Pfaffinger, H. Blumenthal and F. W. Glaser, (90–102). **Titanium-Carbide Products Produced by the Infiltration Technique.** L. P. Skolnick and C. G. Goetzel, (103–110). **Thermal Fatigue Testing of Sheet Metal.** H. E. Lardge, (146–163). Steels were included. **Thermal Shock Testing of High-temperature Metallic Materials.** T. A. Hunter, (164–182). **The Development of Zirconium-rich Protective Coatings and Brazing Materials for Heat-resisting Alloys.** A. Blainey, (183–193).

Steam Tubing for Temperatures Above 550° C. A. Feroci. (*Calore*, 1956, **27,** July, 313–322). [In Italian]. The author discusses the advantages and disadvantages of using austenitic steel tubes for high temperature boilers. The chemical compositions of steels, ferritic and austenitic, and their mechanical properties when hot are examined in detail. Corrosion resistance and welding properties are also discussed. (18 references).—M. D. J. B.

Dynamic Checks Point Way to Longer Chain Life. W. F. Hofmeister and H. Klaucke. (*Iron Age*, 1956, **178,** Aug. 16, 98–99). An investigation of roller chain drives is reported.

Tests showed the effects of sprocket eccentricity, and chain pitch variations. Precision manufacture and careful maintenance are important.—D. L. C. P.

Pre-Setting and Stress Peening Improve Spring Strength. E. H. Spaulding. (*S.A.E.J.*, 1956, Mar., 19–21; *Trans. Soc. Auto. Eng.*, 1956, **64,** 453–457). Increase in life of 50% in springs for limited life is obtained by pre-set deflection to twice the amount producing permanent set and shot peening.

Use of Galvanized Sheets in Concrete Reinforcement. I. A. Benjamin. (*Amer. Iron Steel Inst. Yearbook*, 1955, 347–369). Bond strength measurements are described, also shrinkage and structural performance investigations.

Developments in the Use of Steel in Housing Constructions. J. Dutheil. (*Acier Stahl Steel*, 1956, **21,** Oct., 397–400). A brief review of advantages with illustrations of French buildings.

Galvanized Steel Sheet Roofing. G. Halsband. (*Acier Stahl Steel*, 1956, **21,** Oct., 403–405; from *Continentaler Eisenhandel*, 1956, (4).) A brief description with illustrations of premises.

The Fabrication and Inspection of Pressure Vessels with Special Reference to Those Designed for Sub-zero Conditions. R. Elliott. (*J. S. African Inst. Mech. Eng.*, 1956, **6,** Aug., 1–17). A short description of air receivers made to B.S. 487 is given, mentioning some manufacturing details and also the test procedure required. The more important factors of B.S. 1500 and the classification of vessels are discussed. The fabrication of Class 1 vessels which operate at sub-zero temperatures is dealt with and the method adopted for impact testing the material used at such temperatures is described. The article is followed by an intensive discussion.—L. E. W.

The Construction of Pressure Vessels in 26% Chrome 4–5% Nickel Steels, with or without Additions of Molybdenum. G. Lilliekvist. (*Acciaio Inossid.*, 1956, **23,** Jan., April, 4–14). [In Italian]. The author describes a range of steels developed at Avesta with high chrome and low nickel contents. These are particularly interesting in view of the high cost of nickel and are considered eminently suitable where mechanical strength and high resistance to corrosion are required and where the use of welding can be dispensed with. Details are given of their compositions, structure and mechanical characteristics, their workability and weldability and their resistance to corrosion. The author concludes that they are suitable for many types of pressure vessels.—M. D. J. B.

Design and Construction of Reactor Vessel. G. Brown, M. J. Noone and R. F. Bishop. (*J. Brit. Nuclear Energ. Conf.*, 1956, preprint, pp. 16). The steel and the construction of the reactor shell are described. Pressure and vacuum tests were carried out.

Early Metallurgical Problems. R. A. U. Huddle and L. M. Wyatt. (*J. Brit. Nuclear Energ. Conf.*, 1956, preprint, pp. 20). Early work on the pressure vessel steel is outlined. Most of the paper concerns fuel elements and their canning for nuclear reactors.

Metallurgical Developments. L. Grainger and A. B. McIntosh. (*J. Brit. Nuclear Energ. Conf.*, 1956, preprint, pp. 15). Selection of steel for the Calder pressure vessel is discussed and stress relaxation data are given. Most of the paper concerns fuel elements.

Stainless Steels and High Temperature Alloys in Aeronautics. L. Guitton. *Métaux-Corrosion-Indust.*, 1956, **31,** Oct., 407–417). Typical high temperature steels which are at present in use in gas turbines are considered. The condition of service of various parts of gas turbine are discussed in order to explain the properties of the metal alloys which are required. Creep resistance and resistance to oxidation at high temperatures are the two most important properties of the alloy steels used and the influence of composition on these properties is described. Materials used in British, French and American gas turbines are given.—B. G. B.

A Billion Engine Hours on Aluminium Bearings. R. S. Frank and W. J. Lux. (*Trans. Soc. Auto. Eng.*, 1956, **64,** 655–664). Bonded steel backed aluminium bearings are described. The advantages of these are: (*a*) reduced clearances, (*b*) no life limit below 5000 psi fatigue stress value, (*c*) less sensitivity to high oil temperatures, (*d*) negligible wear (after 29,000 h in one test), (*e*) simpler bearing-locating designs, (*f*) special excellence for high-load high-speed applications.

Application of Metallic Materials for Aircraft Structures in the Temperature Range 600 to 1100° F. J. W. Huffman. (*Trans. Soc. Auto. Eng.*, 1956, **64,** 5–11). Martensitic and austenitic steels, precipitation-hardening and titanium alloys are reviewed and their yield strengths at various temperatures given. Compression, shear, creep and notch effects are briefly considered with notes on fabrication characteristics.

Application of High-Temperature Materials to Aircraft Powerplants in the Temperature Range 1200–2400° C. A. V. Levy. (*Trans. Soc. Auto. Eng.*, 1956, **64,** 366–378). Stainless steels and other alloys, ceramic coatings and cermets are reviewed with physical properties, strength, creep properties, emissivity and cost, and applications to engines of various types.

Structures (Supersonic Flight). D. J. Farrar. (*J. Roy. Aeronaut. Soc.*, 1956, **60,** Nov. 712–720). Properties of steels and other alloys are related to the demands of aircraft structures for supersonic flight.

Tests and Development of Constructions by Use of TOR Steel. S. Soretz. (*PACT*, 1956, **10,** Oct., 411–413). The TOR steels are tested in prestressed ferroconcrete beams.

Methods of Working Chromium-Nickel Steel and Cobalt-Chromium Alloys for Dentures. R. Halbach. (*Techn. Mitt. Krupp*, 1956, **14,** Aug., 60–67). A description is given of the preparation and treatment of Wipla, Witex, Wisil and Wiptam alloys for dental purposes.—L. D. H.

Clinical Experiences in the Use of Chromium-Nickel Steel, and Cobalt-Chromium-Molybdenum and Nickel-Bearing Cobalt-Chromium Alloys in Dental Therapeutics. (*Techn. Mitt. Krupp*, 1956, **14,** Aug., 68–76). Clinical experiences in the use of a number of ferrous and non-ferrous alloys for dental treatment are discussed.—L. D. H.

Materials for the Manufacture of Dentures from Chromium-Nickel Steel, Cobalt-Chromium-Molybdenum and Nickel-Bearing Cobalt-Chromium Alloys. W. Tofaute. (*Techn. Mitt. Krupp*, 1956, **14,** Aug., 54–59). The properties of 18–8 Cr–Ni steel, Co–Cr–Mo and Ni–bearing Co–Cr alloys are discussed, with particular reference to their applications to dental treatment.—L. D. H.

HISTORICAL

Outline of the History of Metallurgy. J. Thomas. (*Mém. Artillerie Fran.*, 1956, **30,** (3), 679–694). A brief account is given with special reference to iron and steel.

The History of Metallurgy—Archeological Metallurgy at the ISML. (*Met. Ital.*, 1956, **48,** June, 185–187). [In Italian]. This article describes the progress made in the history of metallurgy by researches carried out in modern laboratories. Light is thrown on the composition of an Etruscan knife, Etruscan bronzes and spear blades. An indication is given of the probable methods which were adopted to manufacture their implements.—M. D. J. B.

Iron Through the Ages. (*Annales de l'Est, Faculté des Lettres de l'Université de Nancy*, No. 16., 1956, *Report of International Colloquium, Nancy*, Oct., 1955, pp. 592). **Opening Address.** L. Febvre, (17–24). **Evolution of Iron Working in Western Europe from Prehistoric Times to the Middle Ages.** A. France-Lanord, (27–43). **Metallurgical Techniques after the Great Invasions.** E. Salin, (45–56). Metallographic details of objects are given. **Excavations of Sites of Ancient Forges in the Regions of La Sieg, La Lahn and La Dill.** J.-W. Gilles, (57–81). **Iron in Ancient Egypt, the Sudan and Africa.** J. Leclant, (83–91). **Remarks on the History of Iron and Steel Technology in China.** J. Needham. (93–99; in French, 100–108). **Iron and the Iron Industry in European Economics of the XIth to XVIIIth Centuries.** J. Schneider. (111–141). **The Carthusian Ironmasters.** J. Bouchayer, (143–160). The period covered is 1084–1752. **Iron and Steel in the Department of the Meuse from XVIth Century to 1870.** P. Grosdidier, (161–174). **Trade and the Price of Iron in Picardy, XVI–XVIIth Centuries.** P. Goubert, (175–186). **Iron in Classical Swedish Literature.** M. Gravier, (187–197). **The Mazarins, Ironmasters, and Iron and Steel in Alsace under the Ancient Regime.** G. Livet, (199–215). **Influence of the Coalfields on the Location of Ironworks in England before about 1717.** B. L. C. Johnson, (217–223). **Techniques and Civilization Based on Iron in Europe in the XVIIIth Century.** P. Leon, (227–264). **Results of Recent Studies on the History of the Swedish Iron Industry.** M. E. Soederlund, (265–273). **Mines and Metallurgy in the Urals in the XVIIIth Century.** R. Portal, (276–284). **The Brothers John and William Wilkinson and their Contacts with French Metallurgy.** W. H. Chaloner, (285–305). With appendix on William's account of French iron, 1787. **The Significance of the 18th Century in the Evolution of Steel as an Industrial Material.** E. H. Schulz, (303–308). [In German].

The Former Iron Industry of North Périgord in the XVIIIth and Early XIXth Centuries. R. Pijassou. (309–313). **Technical and Human Problems of Contemporary Metallurgy, 1870-1955.** J. Fourastie, (317–343). A brief review with appendices on the output and personnel of the French Industry, prices, U.S. employees and man-hours, World production figures and the effect of progress in the industry on prices of typical products. **The Firm of Gouvy from its Foundation up to the Present. Evolution of Steel-making Processes.** F. Gouvy, (345–353). **The Metallurgy of Alsace. Metallurgical Problems at the Beginning of the XIXth Century (1815–1830).** P. Leuilliot, (355–372). **Notes on the Production of Ploughs in Lorraine at the Beginning of the XIXth Century.** J. -A. Lesourd, (373–382). **A Tradition of Metallurgy in a Rural District. The Firm of Dietrich at and near Niederbronn.** R. Dufraisse, (383–394). **Graffenstaden Works, near Strasbourg, An Example of Evolution and Expansion in the XIXth Century.** F. l'Huillier, (395–410). **The French Iron and Steel Industry at the Beginning of the XIXth Century: Speculations on Iron in 1817.** G. Thuillier, (411–419). **The Role of the Railways in the Transformation of the Forges of the Haute-Marne and Haute-Saône.** O. Voilliard, (421–432). **The Origins and Evolution of Iron and Steel Industry in the Longwy Basins.** M. Grison, (433–451). **The Evolution of the Structure of an Iron and Steel Company of the Central Massif since the Beginning of the XIXth Century: The Commentry-Fourchambault-Decazeville Co.** U. Delaspre, (453–463). **Iron on the [German] Frontier.** H. Contamine, (465–472). Mainly a study of the effects of the 1914–18 war. **Wrought Iron of the Nancy School.** T. Charpentier, (473–478). **Iron Metallurgy in the Mountains: the French Alps.** G. Veyret, (481–490). A study of geographical effects on transport and labour. **Labour Problems of the Lorraine Iron and Steel and Mining Industries.** C. Precheur, (491–502). **The Evolution of Settlement in an Iron Region.** J. Peter, (503–513). An account of growth of population in the Longwy area. **Evolution of Cast Iron in Europe and its Relation to the Walloon Method of Refining.** J. Marechal, (517–533). With bibliography. **Researches on the Trend of Prices in Iron Founding from 1900–1955.** F. Chenique. (With Introduction by J. Donders). (535–553). **Water in Iron and Steel Industry through the Ages.** R. Colas, (555–565). **Closing Address.** E. Salin, (583–588).

Prehistoric Iron in India. (*Edgar Allen News*, 1956, **35,** July, 147). Iron articles found in the cist-graves of Southern India are described. The cairns are among the oldest burial grounds in the world.—D. L. C. P.

The Iron Industry in Milan. F. Peco. (*Met. Ital.*, 1956, **48,** June, 188–200). [In Italian]. An historical review is given of the iron industry in and around Milan from its origins to the present day. Considerable information is given on the latest iron and steel producing installation in the area.—M. D. J. B.

Sir Henry Bessemer (1813–1898). W. H. Chaloner. (*Research*, 1956, **9,** Sept., 323–328).

ECONOMICS AND STATISTICS

Organization of Accounting at the Kuznetsk Combine. N. G. Kropachev and K. A. Popov. (*Stal'*, 1956, (5), 452–459). [In Russian]. The accounting systems used in the ore mines, the sinter and blast-furnace plants, the steel-melting shops, the rolling mills and auxiliary plant are described with special reference to the influence of productivity and quality on payments.—S. K.

Steel's Depreciation Problem. B. F. Fairless. (*Amer. Iron Steel Inst. Yearbook*, 1956, 43–68). The Presidential Address, on mainly economic topics.

MISCELLANEOUS

Investigation of the Surface of a Pyrites Crystal. J. L. Amoros, M. A. Brandoly and M. Perez Peñasco. (*Publicaciones del Dept. de Cristalografia y. Mineralogia*, 1955, **2,** Nov., 89–100). Dislocations and corrosion figures are shown.

Note on Crystals of Iron Aluminide, Al_3Fe. R. Candel Vila. (*Publicaciones del Dept. de Cristalografia y Mineralogia*, 1955, **2,** Nov., 119–122). Crystal parameters are given.

Thermal Decomposition of Manganous and Ferrous Carbonates. H. E. Kissinger, H. F. McMurdie and B. S. Simpson. (*J. Amer. Ceram. Soc.*, 1956, **39,** May, 168–172). The mechanism of decomposition of manganous and ferrous carbonates was explored by high temperature X-ray diffraction in controlled atmospheres, by differential thermal analysis and by thermogravimetric analysis; the results are reported.—D. L. C. P.

A Note on the Ferromagnetic Resonance in α-Fe_2O_3. M. Shimizu. (*J. Phys. Soc. Japan*, 1956, **11,** Oct., 1078–1083).

Magnetic Anisotropy of the Iron Sulphide Single Crystal at the α-Transformation Temperature. T. Kamigaichi, T. Hihara, H. Tazaki, and E. Hirahara. (*J. Phys. Soc. Japan*, 1956, **11,** Oct., 1123).

The Magnetic Properties of $FeSe_x$ with NiAs Structure. T. Hirone and S. Chiba. (*J. Phys. Soc. Japan*, 1956, **11,** June, 666–670).

Electronic Structures of Transition-Metal Complexes. L. E. Orgel. (*J. Chem. Phys.*, 1955, **23,** Oct., 1819–1823).

Metallography of a Space Traveller. C. R. Simcoe. (*Metal Progress*, 1956, **70,** Aug., 72–77). An account of metallurgical examinations of meteorites is given.—B. G. B.

BOOK NOTICES

"*Kempe's Engineers Yearbook*, 1957." 62nd Edition. 2 Vols., 8vo. Illustrated. Volume One, pp. xii + 1324; Volume Two, pp. viii + 1416. London, 1956: Morgan Brothers (Publishers) Ltd. (Price 82*s.* 6*d.* the two volumes, in case).

There are numerous additions to the 62nd edition of "Kempe," mainly, naturally enough, in the engineering fields; notable are a new section on diesel locomotives, soon to make prominent appearance on British Railways, and an up-to-date revision of the section on atomic power taken almost up to the opening of Calder Hall. The metallurgy side contains enlargements of the section on iron and steel, and on foundry practice; there is new text on both magnesium alloys and titanium alloys. The high standards of clarity have been maintained.—J. P. S.

COGHLAN, H. H. "*Notes on Prehistoric and Early Iron in the Old World.*" Including a Metallographic and Metallurgical Examination of Specimens Selected by the Pitt Rivers Museum by courtesy of the Director of Research and Technical Development of Messrs. Stewarts and Lloyds, and Contributions by I. M. Allen, Museum Technician. (Pitt Rivers Museum, University of Oxford, Occasional Papers on Technology, 8, edited by T. K. Penniman and B. M. Blackwood). La. 8vo, pp. 220 + 16 Plates. Oxford, 1956: The University Press. (Price 25*s.*).

Following a monograph on the early metallurgy of the copper and bronze ages—("Notes on the Prehistoric Metallurgy of Copper and Bronze in the Old World," Occasional Papers on Technology 4), the author has now prepared a corresponding account of the earliest history of ironmaking and its gradual evolution. Together, the two form what is possibly the most comprehensive account of the history of our science and technology yet presented. After discussing the ores and early mining techniques, Mr. Coghlan considers the meteoric irons and their fabrication. There is clearly more to learn here than has as yet emerged, and to the non-metallurgist it might have been well to emphasize that the existence in a sample of Neumann lamellæ is not of itself proof of meteoric origin.

The description of the furnaces and the fuels used and their development is most creditably done. Alongside this the description of tools, sometimes surprisingly modern in appearance even when early in date, completes the picture of the techniques and technologies which were employed. One cannot help realizing that these early metallurgists were actuated by a real research sense; the developments can only have occurred as the result of repeated trials and extensive and arduous tests.

The monograph does suffer from slight imperfections evident to the metallurgist. As an example may be quoted the apparent belief (p. 173) that Neumann lamellæ are a normal product of the cold-working of ferrite. It is a little bit doubtful, too, how far the sketches of microstructures will be particularly informative to the non-scientific reader. On the whole, however, these imperfections do not seriously mar the general presentation, but one cannot help remarking on the distinctly impressionistic form of the liquidus and solidus portions of the iron-carbon diagram given on p. 172.

The final portion of this monograph deals with a very able metallographic examination by Mr. P. Whitaker of sixteen samples of iron selected by the Pitt Rivers Museum. Full analyses of many of these are provided and the structures illustrated by most excellent photomicrographs. Much more work of this kind by duly qualified investigators will be required before the full picture can be painted. The Museum, which has published this work, the author, who has clearly gone to such pains to collect and arrange the material, and Mr. Whitaker have done for students of the early history of iron a most valuable service.

F. C. THOMPSON.

POLUSHKIN, E. P. "*Defects and Failures of Metals. Their Origin and Elimination.*" 8vo, pp. xvi + 399. Illustrated. Amsterdam, London, New York, Princeton, 1956: Elsevier Publishing Company. (Price 72*s.*).

The monumental task which the author sets himself is apparent when it is observed that almost all of the fifteen chapter headings have each been the subject of one or more books. The present volume lists 2685 references, which must represent only a small proportion of the relevant literature, and to condense this material into less than 400 pages results in an interpretation which will not meet with universal approval. The author attempts to systematize the whole work by dividing each chapter into six sections, dealing in turn with the characteristics, origin, identification, occurrence, effects and prevention of each defect or failure.

The early chapters deal with defects in cast metals, the treatment being reasonable. The chapter on impurities in metals is biased towards non-metallic inclusions at the expense of soluble impurities and the following one on the decarburization of steel is adequate in the circumstances. The treatment of scaling is sketchy and short, having regard to the importance of the subject, and the chapter on residual stresses leaves much to be desired, not only in the manner of its presentation but also in the validity of some of the views expressed; there is already much confused thought in this field, and the author will not help to resolve this confusion by reference to "plastic micro-stresses" and autofrettage as a method of stress-relief, while virtually ignoring the question of residual stresses in welded structures and castings. The next chapter, on the fatigue of metals, is inadequate for the most important mode of mechanical failure, although the treatment of flakes and failures in heat-treatment is of a higher standard. The author hardly does justice to his remark in the Preface "The current nomenclature is not entirely satisfactory, some terms of common usage are not sufficiently explicit and may be misleading." when, in the chapter on the embrittlement of metals, he perpetuates the term "caustic embrittlement," while pointing out that the plate material involved is quite ductile. In the section on the origin of cracks there is, of necessity, some overlapping with the contents of other chapters, and that on defects in shape, surface and size of metals reads like a dictionary. The treatment of wear of metals with only a passing reference to the work of Bowden and his school is amazing, while the subject of corrosion, even in 66 pages, sets quite a problem to any author.

It is obviously easy to be critical of a book which attempts to cover so much ground as the present one by reference to its omissions, of which creep is one in this case. However, when space is at such a premium clearly it should not be wasted, and this is perhaps the most important single criticism which can be levelled at the present volume. The style of writing does not conserve space, and should a second edition be called for the author may consider avoiding sentences such as those in the eight lines devoted to the corrosion of low-carbon steel, where three lines are taken up by: "Three stages of corrosion were observed. In the first stage the metal displayed only initial attack, in the second it became progressively covered with rust, and in the third rust covered the entire surface."

The book is well produced and contains many excellent photographs. Its greatest use will be as a source book to those actively engaged in the art of detection involved in solving metal-failure mysteries.—R. N. PARKINS.

BOWDEN, F. P. and D. TABOR. "*Friction and Lubrication.*" Sma. 8vo, pp. x + 150. Illustrated. London, 1956: Methuen & Co., Ltd.; New York: John Wiley & Sons, Inc. (Price 10*s.* 6*d.*).

The essence of the Authors' account of the nature of friction is that, when even the smoothest of surfaces are placed together, contact occurs only at the tips of asperities. The load being concentrated at a few small points, high local stress occurs and the material yields so as to share the load more widely. This continues until the area of contact, when multiplied by the plastic yield stress of the material equals the applied load. Strong junctions occur at points of intimate contact, and the force of friction is considered by the authors to be that required to break the junctions In experiments on outgassed iron sliding on outgassed iron in a vacuum one part seizes firmly on the other. When, however, oxygen is admitted the coefficient of friction falls immediately to a value of 2 and continues to fall until, after 15 hours, the normally quoted value of 0·5 is attained. The oxide formed on a metal is thus of considerable importance, and the nature of that oxide will affect frictional behaviour between rubbing bodies.

Lubrication may be divided into two types, "hydrodynamic," wherein sufficient pressure is generated within a fluid to separate the surfaces completely, and "boundary" wherein the chemical properties of the lubricant are important. Sir William Hardy considered that boundary lubricants were effective because they were physically absorbed on to the surface. The Authors adduce evidence to show that the fatty acids, for example, react with material of the surface to form soaps. The effectiveness of lubrication is then dependent on the rheological properties of the soap.

Because true contact between rubbing bodies is concentrated over a minute area, the energy dissipated by friction will be confined to a small quantity of matter so that high local temperatures may be expected. Such local hot spots may initiate explosions.

The book, which clearly states the authors' viewpoint without mathematics, is well illustrated. The modern physical methods by which the authors have reached the above conclusions are described and, additionally, chapters are devoted to rolling friction, friction of non-metallic solids and extreme pressure lubrication. Whilst the basis of the book is academic rather than technological, the clear account of fundamental information which it gives will provide an indispensible foundation for the understanding and solution of lubrication problems which occur in practice.

F. T. BARWELL.

"*Réaumur's Memoirs on Steel and Iron.*" A translation from the Original printed in 1722. By Anneliese Grünhaldt Sisco, with an Introduction and Notes by Cyril Stanley Smith. La. 8vo, pp. xxxvi + 396. Illustrated. Chicago, Ill., 1956: The University of Chicago Press. (Price $6·00).

To most of us Réaumur's name is recalled only in connection with a (possibly obsolete) scale of temperature and a vague idea that he was responsible for the invention of the production of malleable cast iron. These, however, represented but a small part of his scientific activities, and it may well be that his work on insects is the one by which he will be most remembered. It is not, also, without interest that he may be regarded as the ultimate ancestor of the wood-pulp paper industry. Réaumur was a typical product of his time. To him the discovery of new knowledge and its application to the needs of the community were inextricably linked together. As Professor C. S. Smith says in his most illuminating introduction, "he was, in fact, one of the first writers on any topic who can be called an applied scientist in the modern sense."

Although only four out of the seventy-five publications listed in the Introduction deal with metals, these are the ones by which he will be remembered by us. They are of outstanding importance, and, quite apart from his work on malleable cast iron, he did much on the production of tinplate. The work was, however, no mere *ad hoc* investigation of a problem or problems which needed solution for utilitarian processes. He was a real scientist, though possibly not an outstanding one, and his approach differed in no essential from the scientific approach of the present day. It is, in fact, astonishing to note how closely the general pattern of his thought coincides with that of the most modern research worker. His interest in diffusion, for example, led him to suggest that "diffusion could produce tiny voids and that porous solids would become more dense on prolonged annealing." The metal physicists

concerned with vacant sites in the lattice will, therefore, immediately find a common link between their work and his.

By sponsoring the translation of these two books—for such the two parts of this volume really are, the first dealing with "The Art of Converting Wrought Iron into Steel," and the second with "The Art of Making Cast Iron Malleable," the American Iron and Steel Institute has laid all interested in the history of our science and technology under another debt of gratitude. The translation is excellent and Professor Smith's introduction quite first-rate. To all who have in any way contributed to the production of this most interesting volume our thanks are due.

F. C. THOMPSON.

NEW PUBLICATIONS

AMERICAN SOCIETY FOR TESTING MATERIALS. "*ASTM Standards on Zinc-coated Iron and Steel Products.*" (Zinc-coated Sheets, Hardware, Wire, Strands, Fencing and Pipe, Terne-Alloy-Coated Sheets). La. 8vo, pp. viii + 134. Illustrated. Philadelphia, 1956: The Society. (Price $2·25).

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Report on Standard Samples and Related Materials for Spectrochemical Analysis, 1955.*" Compiled by Robert E. Michaelis, National Bureau of Standards, for ASTM Committee E-2 on Emission Spectroscopy. (ASTM Special Technical Publication No. 58–C.). 8vo, pp. v + 87. Philadelphia, Pa., 1956: The Society. (Price $2·75).

BERG, HANS-JOACHIM VAN DEN. "100 *Jahre Friedrich Siemens Regeneratio-Öfen.* 8vo, pp. 77. Illustrated. Berlin, 1956: Friedrich Siemens GmbH.

BRITISH STANDARDS INSTITUTION. B.S. 18 : 1956. "*Tensile Testing of Metals.*" 8vo. pp. 24. London: The Institution. (Price 4*s.*).

BRITISH STANDARDS INSTITUTION. B.S. 215 : Part 2 : 1956. "*Steel-Cored Aluminium Conductors for Overhead Power Transmission Purposes.*" 8vo, pp. 20. London: The Institution. (Price 3*s.* 6*d.*).

BRITISH STANDARDS INSTITUTION. B.S. 2803 : 1956. "*Oil Hardened and Tempered Steel Wire for Springs.*" 8vo, pp. 11. London: The Institution. (Price 3*s.*).

BROWN, B. F. "*Effect of Baking on Delayed Fracture of Electroplated Ultra High-Strength Steel.*" (Naval Research Laboratory Report 4839.) La. 4to, pp. ii + 18. Illustrated. Washington, D.C., 1956: Naval Research Laboratory.

CHAMBRE SYNDICALE DES PRODUCTEURS D'ACIERS FINS ET SPÉCIAUX. "*Journée des Aciers Spéciaux, le 30 Novembre* 1956." (IVe Salon de la Chemie, 22 Novembre au 3 Décembre, 1956.) La.4to, pp. 163. Illustrated. Paris, 1956: Chambere Syndicale des Producteurs d'Aciers Fins et Spéciaux. (Mimeographed).

CLARK, M. GARDNER. "*The Economics of Soviet Steel.*" (Russian Research Centre Studies No. 25.) La. 8vo, p. xvi + 400 with map. Cambridge, Mass., 1956: Harvard University Press. (Price 60*s.*).

DELBART, G. and A. CONSTANT. "*Courbes de Transformation des Aciers de Fabrication Française.*" Vol. II. La. 8vo, (loose-leaf). Illustrated. Saint Germain-en-Laye [1956]: Institut de Recherches de la Sidérurgie.

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A. H. INGEN-HOUSZ

ARNOLD HUGO INGEN-HOUSZ was born at The Hague. He received his technical education at the Technical University of Delft, where he obtained a degree in mechanical engineering in 1911. His first appointment was with Royal Dutch Shell; he worked chiefly in Eastern Europe (Rumania and Russia) until 1914, when he returned to The Hague as engineer of its municipal gasworks. In 1917 Mr. Ingen-Housz joined the Royal Netherlands Blast Furnaces and Steel Works Ltd. (Koninkl. Nederlandsche Hoogovens en Staalfabrieken N.V.), as technical assistant to the founder of the company, Mr. Wenckebach, in the planning and construction of the IJmuiden plant.

On the death of Mr. Wenckebach in 1924, Mr. Ingen-Housz was appointed Joint Managing Director of the Company, together with Mr. Kessler, who had been assistant to Mr. Wenckebach for economic matters. He became President of the Management Board (Directie) in 1945 on the death of Mr. Kessler, a position which he still occupies. He is also President of the Management Board of Breedband (wide strip mill and tinplate plant) and a member of the Boards of Mekog (nitrogenous fertilizer plant) and Cemij (cement plant); all three companies are affiliated to the Royal Netherlands Blast Furnaces and Steel Works Ltd.

Mr. Ingen-Housz plays a prominent part in Dutch industrial life; he is a member of the Boards of the Federation of Netherlands Industries and of the Central Social Federation of Employers. He is President of the Association of the Steel Producing Industry, and represents the Netherlands steel industry as a member of the Consultative Committee of the European Coal and Steel Community. He also takes a great interest in education and is a guardian of both the Municipal University of Amsterdam and the Rotterdam School of Economics.

When The Iron and Steel Institute held a Special Meeting in the Netherlands in 1953, Mr. Ingen-Housz, a Member since 1931, was Chairman of the Reception Committee, and in recognition of his work he was elected an Honorary Vice-President of the Institute. He was elected in 1957 to be President of the Institute for 1957-1958, the first overseas Member to receive this distinction since the late C. P. E. Schneider, who served from 1918 to 1920.

A. H. Ingen-Housz

President, 1957–1958

PRESIDENTIAL ADDRESS

The Iron and Steel Institute

The Birth of an Industry

By A. H. Ingen-Housz

I should like to begin by offering my sincere thanks for the honour and distinction you have accorded me in giving me, a foreigner, the opportunity of holding this distinguished office as your President.

I am the more conscious of this honour when I remember that it is 40 years since the last foreigner occupied this position—and especially when I recall who he was. For your President at that time was Mr. Eugène Schneider, the leader of the world-famous Schneider-Creusot concern, an old-style *maître de forge* and a man of European renown. to whose memory I wish, on this occasion, to pay respectful homage.

In contrast, I am a citizen of a small country, which is industrially still young; our Netherlands iron and steel industry, at any rate, was founded only very recently. It is, indeed, almost exactly 40 years ago that the founder of our industry made known his plans to a few interested persons; as a result, these plans were further elaborated, leading ultimately to the erection of the Royal Netherlands Blast Furnaces and Steel Works Ltd. at IJmuiden.

It has occurred to me that it might be of interest to you, as members of an industry which has been in existence since time immemorial, to hear how, in these modern times, a young industry has grown up—and how it established its position in the world. This development has once again proved that the iron and steel industry is no longer a strictly national industry; it is rather an international profession which finds points of contact in other countries in all the problems of technique, trade, and management, and which reveals to an increasing extent the same characteristics over the whole world. This fact has long been demonstrated by the very large overseas membership of The Iron and Steel Institute, which today is being symbolized by my taking the Chair.

I HAVE ALREADY MENTIONED that the iron and steel industry in the Netherlands is a young industry, although in the 18th and 19th centuries there did exist in the eastern part of my country a number of small charcoal blast-furnaces which made cast-iron products using the local bog-ore. However, these furnaces fell into disuse as iron producers when the supplies of charcoal were exhausted, although the factories continued to be used as iron foundries using pig iron bought elsewhere.

A reason why, even around the middle of the 19th century, an iron and steel industry could not be developed, may be the fact that in those days metal-working in the Netherlands was of relatively small importance. The impetus for the growth of a metal-working industry was largely a consequence of changes in ship design, involving the building of steel ships, the manufacture of ships' engines, and the construction of harbour installations. When the scale of metal-working began to increase in this

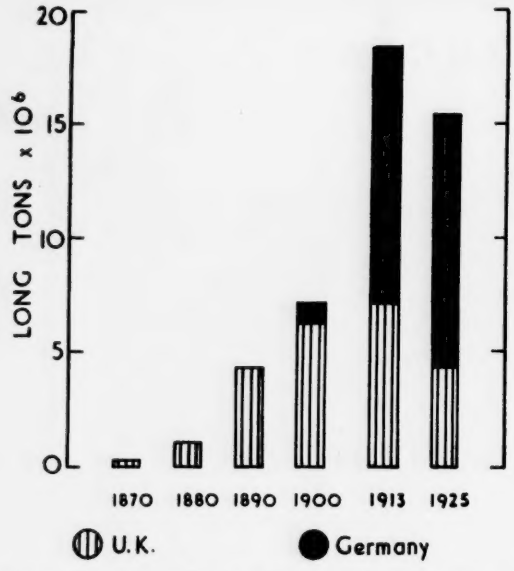


Fig. 1—Imports of iron ore by the U.K. and Germany

way, there were already in existence in neighbouring countries, such as the United Kingdom, Belgium, and Germany, flourishing and extensive iron and steel industries, which could provide the steel needed by the Netherlands. By the beginning of the 20th century, the use of steel in the Netherlands had increased to such a degree that the establishment of a native steel industry became possible. Yet the seemingly unassailable position of the foreign steel industries, which were literally based on the raw materials ore and coal, usually caused the vague dream of a Dutch steel industry to fade quickly.

For the establishment of an iron and steel enterprise in the Netherlands another factor was, however, of much greater importance. I have in mind here the structural changes in the industry which arose in both the United Kingdom and Germany, caused by the fact that these old iron and steel producing countries had to import ever larger quantities of iron ore from abroad (Fig. 1).

The stream of foreign ore for Germany was largely transported over the cheapest supply route for the Ruhr area, the Rhine, and therefore it passed through the Netherlands' port of Rotterdam, which gradually became an important transhipment centre for ore. For example, the supply of ore via Rotterdam, of which there was none in 1870, had risen to 8 million tons* in 1913. Then, too, the Netherlands was centrally located at a point where there was international competition in the world trade in coal, principally from the United Kingdom and Germany. By this time the production of coal had also begun in the Netherlands itself.

Another factor which made the plan for the establishment of a steel industry in the Netherlands particularly attractive was the extensive and well-kept network of canals and rivers. This would make it possible for products to penetrate 500 miles into the interior via the Rhine to the port of Basle, and by canal to reach Paris and the whole of Northern France.

The man who took the initiative and founded the iron and steel industry in the Netherlands, Mr. Wenckebach,† had therefore, after detailed study, come to the conclusion that an enterprise located near the sea could compete with the older industries, because it would be able to receive the raw materials at its own seaport and distribute its products cheaply along waterways. Moreover, this advantage in the matter of transport costs would increase with the years; the old-established enterprises would have to face increasing exhaustion of their raw-material resources, and would thus become committed to the more expensive solution of supply by rail or, if from overseas, by repeated transhipment.

The changes in the supply of raw materials for both the Continent and the United Kingdom which have taken place since that time have proved that this development was correctly foreseen at the time the industry was established at IJmuiden. There are only a few favoured spots left in the world where ores and coal are found close to one another, such as Corby and Alabama. Already the United Kingdom and the continental steel industry are calling to an increasing extent on ores imported from overseas. From Fig. 2 it seems clear, moreover, that this tendency will continue in the future.

Another advantage of a site on the coast which can be mentioned here is that one is not dependent on any particular source of raw materials. For instance, if a Spanish ore mine becomes exhausted this does not constitute a serious problem to us, because we can gradually switch over to ore mines elsewhere.

It is to be expected that in the future the ore will have to be carried from territories located farther and farther away and that this will eventually make the problem of the cheapest possible supply particularly acute.

In this respect, the British Iron and Steel Research Association, under the direction of Sir Charles Goodeve, has in my opinion done epoch-making work by carrying out fundamental studies into both the ships and the unloading installations which must be used for this supply. These are particularly important, because as the distances to be covered become longer and the ore carriers larger, they can be handled in only a limited number of ports. Possibly the construction of large central transhipment installations where the cargoes of large vessels may be transhipped or stored will help to solve the problem. An installation of this kind is now being built in the port of Rotterdam. From it, ores brought over the ocean in large ships can be sent on in smaller vessels to the United Kingdom or in lighters to the blast-furnaces of the Continent. Such a development will, on the whole, make location on the coast of even greater importance.

For this reason, it is not surprising that a number of new steel concerns have been built or are planned

*Throughout this address, ' tons ' mean ' metric tons.'

†H. J. E. Wenckebach, born at The Hague, 27th June, 1861, died 22nd February, 1924.

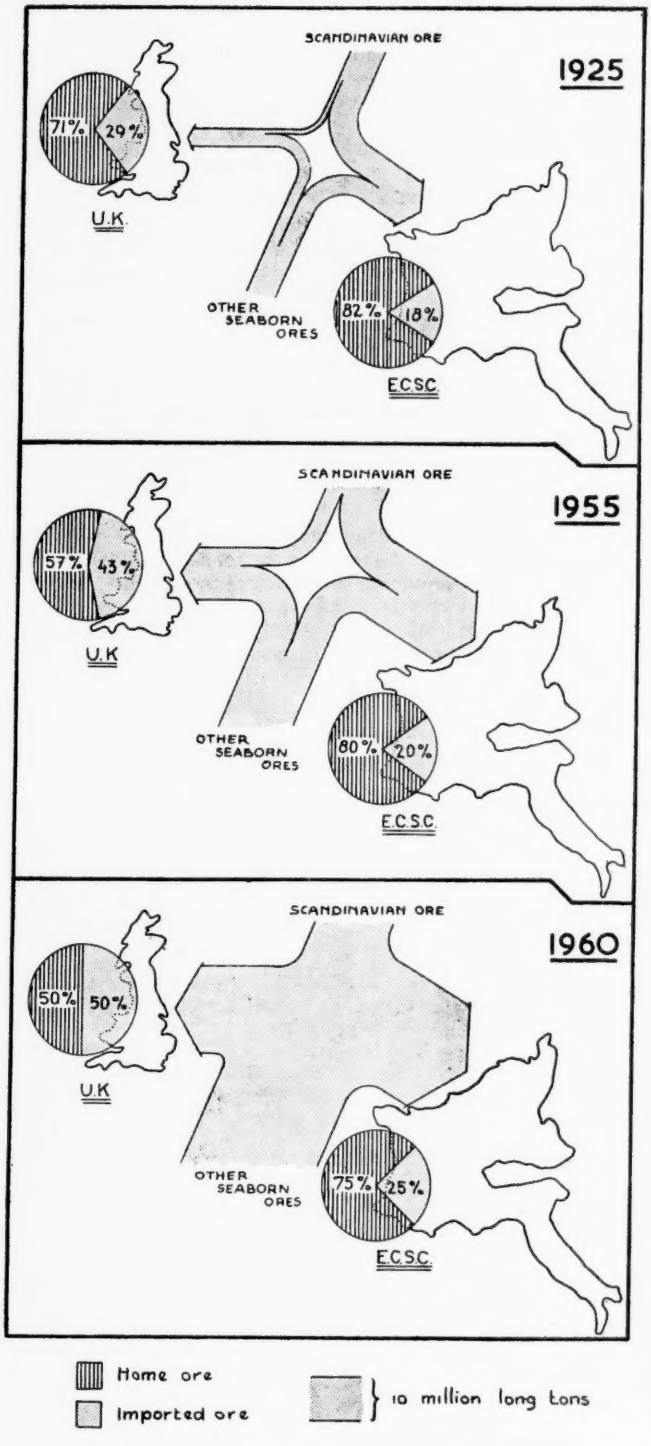


Fig. 2—European iron-ore supplies, 1925, 1955, and 1960

near the sea. In the United Kingdom the enterprise at Port Talbot is known to you all. In Germany the steel industry at Bremen is in full development. In France, in addition to the well-known works at Mondeville, the newly planned establishment at Dunkirk may be mentioned. In Italy, the steelworks at Cornigliano lies on the coast, and the construction of a works at Savona has been announced. Finally, in the United States, too, a similar development can be observed, as shown by the well-known Fairless plant on the Delaware.

Having sketched this general outline, I now return to the development of our own Dutch steel industry, the preparations for which started in 1917. And here I would point out that its establishment was not the result of any nationalistic ambition for self-sufficiency born during the first World War, however understandable such an ambition might have been. The structural changes which I have mentioned had already taken place before 1914, so that implementation of the plans was retarded rather than promoted by the outbreak of war, because only after the armistice was it possible to learn of the recent technical developments in other countries.

The aim of the founders of the Netherlands enterprise was to create a fully integrated iron- and steelworks. Since the execution of such a vast plan soon appeared to be achievable in stages only, it was decided that as a first phase a merchant blast-furnace undertaking should suffice. It was obvious that for the establishment of a steel industry in a country which had not had one before, the technique of steel manufacture abroad had to be studied. We approached our task with open minds, without any traditional preconceptions, without any commitments to national equipment suppliers, and with a ready eye for the merits and defects of the current opinions in foreign iron- and steel-producing concerns.

The decision that for the time being the construction of a blast-furnace plant would have to suffice, naturally meant that, despite the modest initial set-up, full provisions should be made in the plans for those future lines of development which could be foreseen at that time and which would be financially possible. A careful study of blast-furnace plants in various countries, particularly those in Belgium, the Ruhr area, the United Kingdom, and the United States, led us to adopt the U.S. set-up. Even at that time (1920) this made wide use of standardization in its main features. The following character-

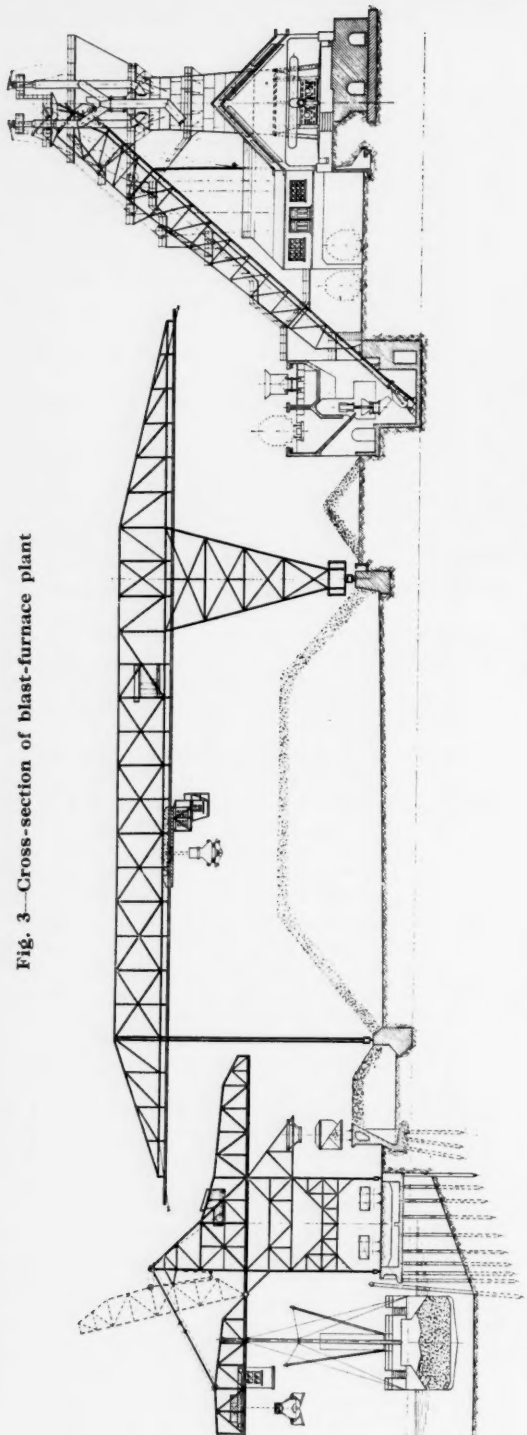


Fig. 3—Cross-section of blast-furnace plant

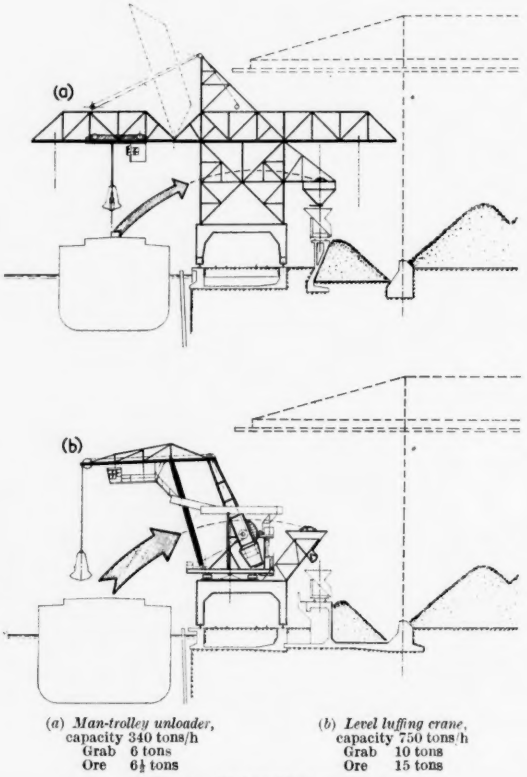
istics in particular were attractive to us:

(*a*) Logical and economic transport and storage of the raw materials between harbour and blast-furnace bunkers, especially in discharging ocean-going vessels
(*b*) Blast-furnace charging by means of a skip-hoist
(*c*) The use of an armoured shaft as a support for the top
(*d*) The consequently lower building costs of the U.S. project.

If we now re-examine the decision taken at that time, two facts are remarkable for having largely made possible the present state of the undertaking.

The first point to be mentioned is that neither the process for the manufacture of pig iron nor the U.S. blast-furnace and auxiliary plant has fundamentally changed during the past four decades. The units have, of course, become considerably larger and the process has been improved by, for example, high top-pressure, the use of oxygen, and ore preparation, although even now these practices have not yet won general acceptance.

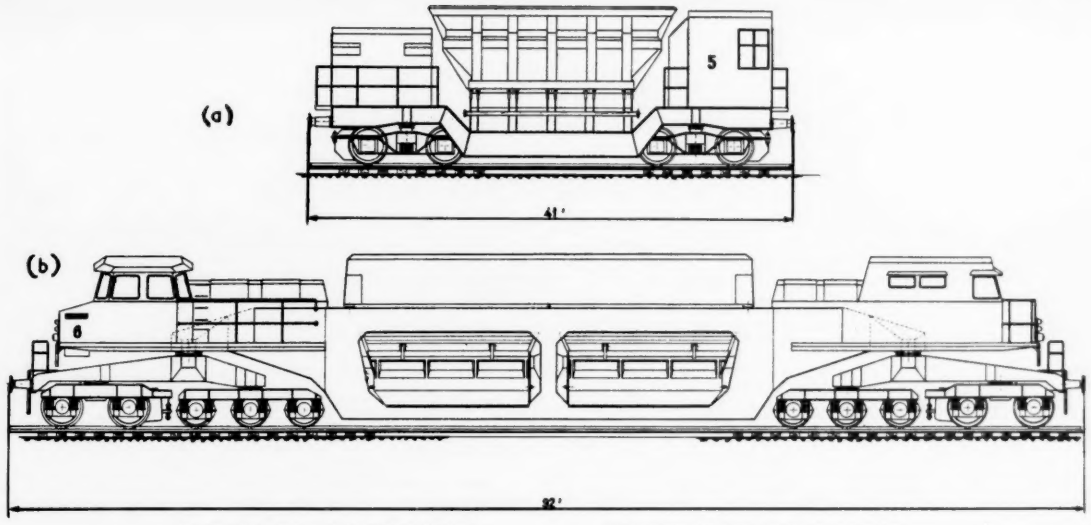
Since an adjoining site was available for ore preparation, the layout chosen by our founder in 1920 proved to be adaptable, without major obstacles, to the desired larger production, provided that arrangements could be made for the greatly increased movement of materials connected with the enlarged capacity.



(*a*) *Man-trolley unloader*, capacity 340 tons/h
Grab 6 tons
Ore 6½ tons

(*b*) *Level luffing crane*, capacity 750 tons/h
Grab 10 tons
Ore 15 tons

Fig. 4—Unloading cranes



(*a*) Capacity 70 tons, total weight 110 tons (*b*) Capacity 160 tons, total weight 290 tons

Fig. 5—Hopper scale cars

The transport arrangements which proved necessary to feed raw materials not only to the original three 15-ft furnaces, but also to a new fourth 28-ft blast-furnace, will certainly be of interest to the engineer and are therefore illustrated here with a number of diagrams.

A cross-section of our blast-furnace plant (Fig. 3) clearly shows the usual U.S. pattern of a quay with separate unloading cranes; the self-propelled unloaders take care of the disposal of the ore. In this respect it should be borne in mind that, owing to the many grades of pig iron which we manufacture, the varieties of ore which we use number more than 40! Each unloader has been fitted with a weighing installation, and accounts for the purchase of ore are settled on the basis of this assessment of weight.

The ore bridges carry the ore from the trough, usually via stockpiles in the ore yard, into the transfer car, which moves along the whole length of the stockyard and keeps the blast-furnace bunkers filled. These bunkers need only have the relatively small capacity of one or two days' supply to cover possible interruptions in the preceding phases.

Finally, there is the scale car under the bunkers which feeds the load into the skip-hoist.

The larger volume of materials to be shifted from harbour to blast-furnace, as well as the increasing size of the ships, made it necessary, or at any rate desirable, to increase the capacity of the unloading cranes. This was done by replacing the unloader-type cranes with level luffing cranes (Fig. 4). The weight on the existing quay could be kept within acceptable limits despite the doubling of the gross load. It might also be mentioned that through this changeover the free digging capacity was increased from 340 to 750 tons/h.

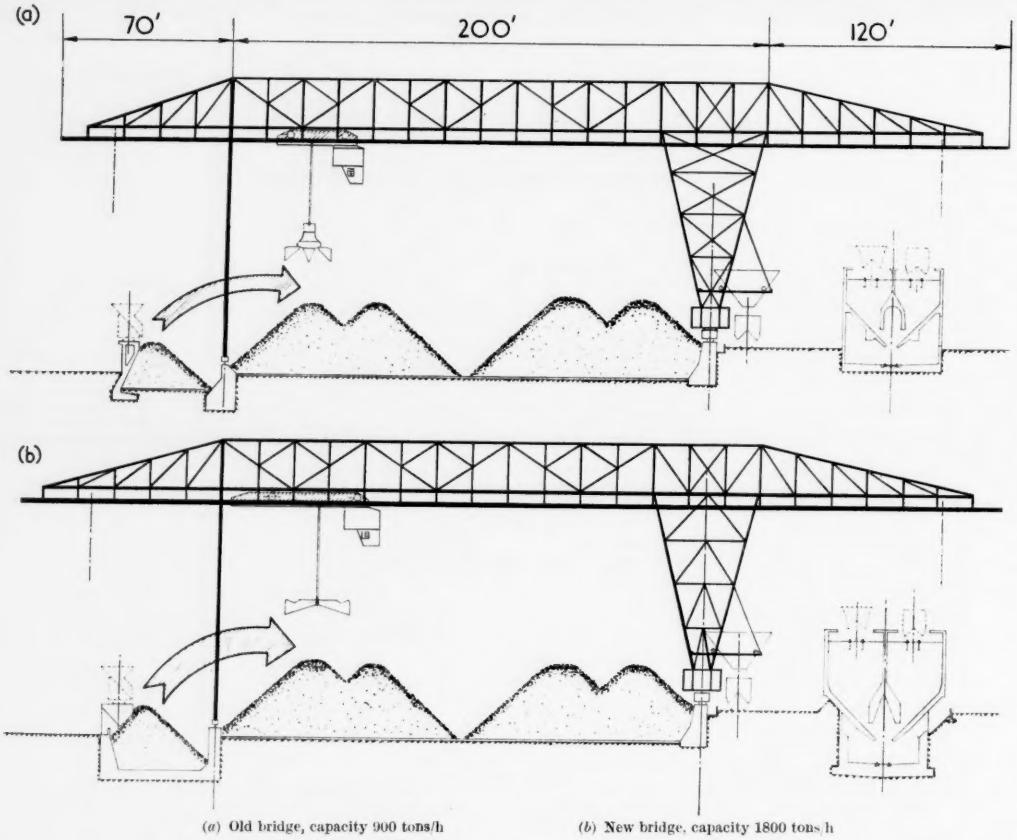
The larger unloading cranes, however, demand faster disposal of the ore by the hopper scale car. Here we were faced with the problem that only one track was available and that the capacity had to be enlarged without increasing the number of cars. This led to the conception of a hopper scale car of exceptional size (160 tons capacity and 92 ft in length), which is now on order and which differs very considerably from the 70-ton cars which have been used so far (Fig. 5).

The means of transport next in the line, the ore bridges, also had to be enlarged in order to empty the trough sufficiently quickly, although in this respect, too, the desired result had to be achieved with the existing crane runway. We have succeeded in doubling the capacity (Fig. 6), even with the same bridge structure, by a more equal distribution and spreading of the load of the trolley wheels, by application of Ward–Leonard and Amplidyne control of the motors, by building the cabin and parts of the trolley in aluminium, and by choosing a grab with a wider spread and a higher degree of filling. Figure 6 shows the increase of the capacity from 900 to the present 1800 tons/h.

With regard to general transport we have changed from hopper cars of 60 tons capacity to cars of 100 tons capacity (Fig. 7), while retaining the 84 lb/yd internal track.

Finally, as far as I know, our mixer-type hot-metal torpedo cars are also the largest in use in Europe. Here, too, the load has been spread as much as possible over a great number of axles (Fig. 8). These cars require 128-lb rails.

As a result of this expansion of the transport capacity which, in view of the special requirements, we have largely designed ourselves, it has proved possible to increase the pig-iron capacity, which in 1920 was forecast as 250,000 tons per year for two blast-furnaces, to about 700,000 tons per year, using three blast-furnaces; with the existing transport capacity a pig-iron production of 1,500,000 tons per year could be handled.



(*a*) Old bridge, capacity 900 tons/h

	Old	*New*
Total load	85 tons	108 tons
Trolley	60 tons	68 tons

(*b*) New bridge, capacity 1800 tons/h

	Old	*New*
Grab ore	25 tons	40 tons
Grab	13 tons	16 tons
Ore	12 tons	24 tons

Fig. 6—Ore bridges

The efficient conception of the U.S. blast-furnace plant has certainly found more and more recognition, and it is, therefore, understandable that this system is being followed in many countries, including the United Kingdom, Russia, and also, to a certain extent, Germany.

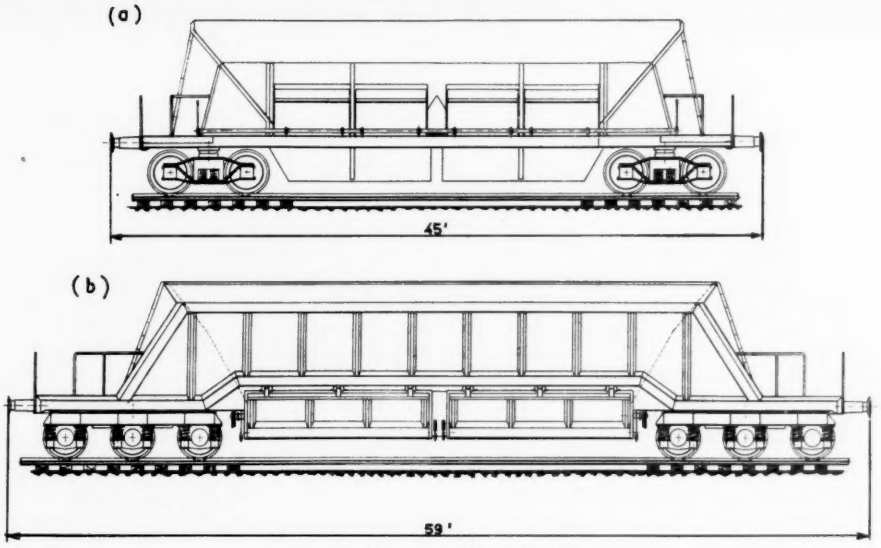
The second point I would like to emphasize in connection with our original choice is that it is quite possible that, even if in 1920 the money had been available to build steelworks and rolling mills at the same time, it would have been found later that what then had been considered a reasonable and sufficiently modern solution for a limited market would have proved obsolete before the second World War, or at least shortly after that war, with the result that we should very probably have been compelled to make drastic changes in the layout and siting of the plants.

Figure 9 shows the present buildings on our site. The modest beginnings in 1924 are marked in black; all later extensions are mainly grouped in parallel strips. The steelworks and rolling-mills as planned in 1920 (dotted lines) would have seriously impeded later developments, not only by reason of their internal design, but also, as Fig. 9 shows, by their siting in the overall plan.

I mention these facts because this actual example clearly shows that, although a proposal may appear to be technically and economically justified and to guarantee, as far as is possible, its financial soundness, the planners of a long-term project are dependent on unpredictable factors. The advantages and disadvantages of these cannot be avoided once the original decisions have been taken.

Had it not been for the relative stability over several decades of the U.S. system of pig-iron manufacture, and had we not been reluctantly compelled to renounce the construction of a steelworks with rolling mills, the phased growth of our concern could not have been accomplished in such a satisfactory way. Partly because of these favourable factors, it has been possible to adapt the undertaking to meet the growing domestic need for steel and to take an increasing share in the export market. In its total conception, including the older part, the works has remained sufficiently modern to be competitive.

The realization that with the same foresight even the development in IJmuiden might easily have



(*a*) Capacity 60 tons (*b*) Capacity 100 tons

Fig. 7—Hopper cars

been much more unfavourable adds to our understanding of the difficult position in which many British iron and steelworks found themselves during the years 1920 to 1935. In a country where the iron and steel industry was established so much earlier, where several companies had chosen their location near ore and coking-coal deposits or where the local market could easily be served, and where the lay-out of many undertakings was only modest, with little room for expansion, it must have cost much thought and sacrifice to carry out modernization, progressive integration, and adaptation to the use of imported ores. Furthermore, I fully realize that other factors also have played their part in the U.K., and that, for instance, social difficulties have come to the fore. Nevertheless, even before the second World War it was possible in the U.K. to erect some modern plants, for instance Corby, Cardiff, and Ebbw Vale, each well adapted to the local conditions. This development, which was continued after the war, for example, with Port Talbot, fills me with great respect for the capability of the British iron and steel men.

Coming back to the development of my own concern, I must admit that the fundamental decision that in the first phase a merchant blast-furnace works would have to suffice involved a particularly great risk. For it signified that we were thus entering into competition with the famous British industry, which sold its pig iron over the whole world, principally by means of easily negotiable warrants. When in 1924 our first pig iron came on the market, we were therefore convinced that we should not be able to sell our products as a mass-produced article, as had been customary until that time, but should have to find new techniques of selling. Our company therefore began immediately to organize sales in another way, by appointing its own representatives and by introducing the pig iron as special brands, so that emphasis could be laid on the good quality of the iron. Another device was to adapt the various grades to the special needs of the customer, after close consultation so as to ensure the correct choice in the light of his particular requirements.

Our company, briefly known as 'Hoogovens', therefore became a merchant-furnace undertaking with a great range of pig-iron varieties and with customers spread over the whole world; this method of operation was, of course, helped by the location of the works on the sea and on the great Continental waterways.

Once our pig-iron production had become a technical and economic success, attention was not immedi-

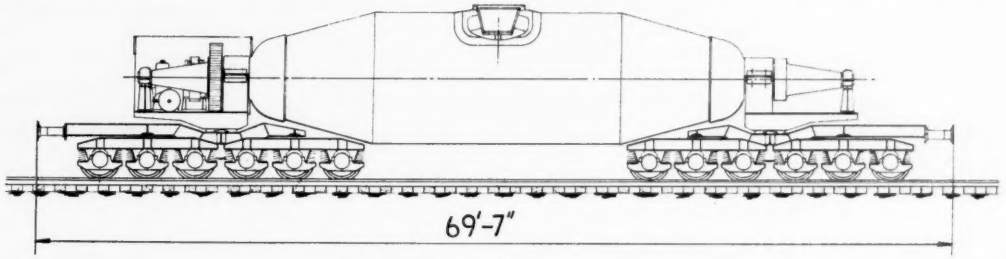


Fig. 8—Mixer-type hot-metal car

ately directed to the manufacture of steel, but to the processing of by-products, through which the cost of the pig iron could be made as low as possible. A supply contract for the utilization of the surplus blast-furnace gas was therefore concluded with the provincial electricity station. In addition, contracts were concluded for the supply of coke-oven gas to municipal gas undertakings, whilst plans were also made for making nitrogenous fertilizers from coke-oven gas and cement from blast-furnace slag.

Finance was a difficulty in this respect. The company was in a very early stage of development and had not yet succeeded in building up sufficiently large reserves nor sufficient credit to be able to finance these big projects itself. However, when partners were found who were also interested in these products, two subsidiary companies were established at the Hoogovens site. Both of these are joint undertakings with other companies. These joint undertakings receive their raw materials from Hoogovens and are under the same centralized management, so that their operation is closely integrated with that of the main concern. In this way a synthetic ammonia plant and a blast-furnace cement works have come into being. The former, which has been considerably extended during the last 25 years, now produces more than 500,000 tons of nitrogenous fertilizers a year. The annual production of the cement factory will also shortly pass the half-million tons mark.

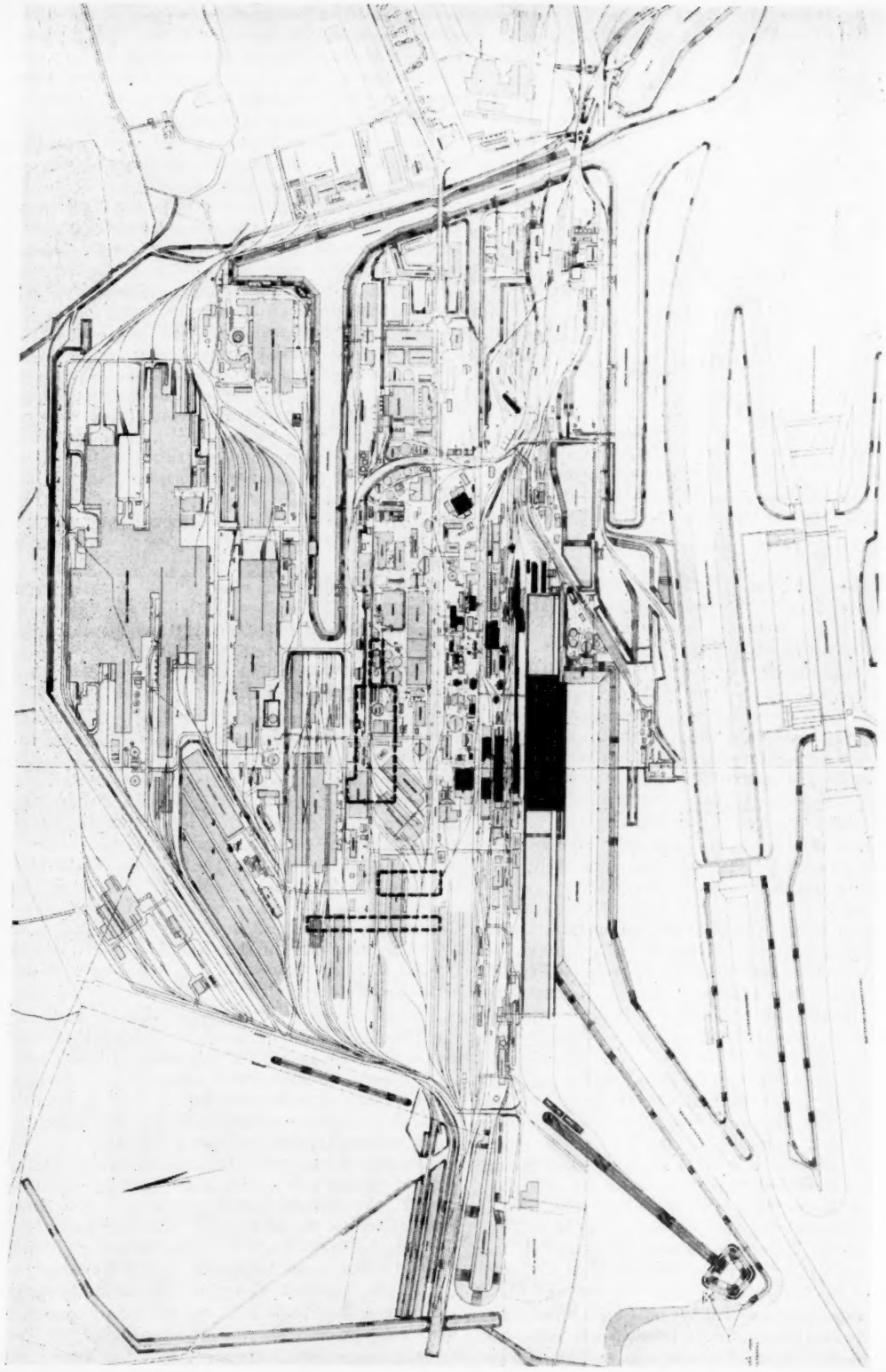
The building of these branch industries had scarcely been completed when the great world economic crisis erupted in full force in 1930. As you will know, this period was a very difficult one for the steel industry all over the world. In the U.K. a solution was found by imposing a high import duty which provided a protective wall behind which the industry could begin modernizing itself. But this solution meant that the situation in other countries, which saw their exports suddenly cut off, became more difficult. The young blast-furnace plant in IJmuiden, owing to the limited market in its own country, was of necessity committed to exports, and so it was threatened in this period with the progressive loss of its newly won markets. In particular, the sudden imposition in the U.K. of the $33\frac{1}{3}\%$ import duty in the spring of 1932 made it necessary for us to apply all our inventiveness to finding new market possibilities. That we succeeded in doing so is shown by Fig. 10.

During the depression the importance of a coastal site, through which a switch to other markets could quickly be effected, became strikingly clear. This was further accentuated by the fact that during the years of crisis transport costs decreased sharply. I remember a freight comparison from this period according to which the transport of one ton of pig iron by ocean-going vessel to the East Coast of the U.S.A. cost $1, precisely the same as transport by river-vessel from IJmuiden to Paris or by rail from IJmuiden to the centre of our country (50 miles, or about the distance from London to Cambridge). In addition to the sale of by-products to which I have already referred, it was partly as a result of these low freight rates that it proved possible to tide over the period of crisis without any appreciable decrease in pig-iron production, although earnings reached their lowest figure.

As soon as economic conditions in the steel industry began to improve, a start was made with the realization of the original plan to complete the undertaking with a steelworks and rolling mills. The decision about the quality of steel to be manufactured was not particularly difficult, since at that time the choice was limited to basic Bessemer or open-hearth steel. An advantage of open-hearth steel production was the fact that overseas ores are preponderantly poor in phosphorus content, and so open-hearth steel can be manufactured relatively more cheaply on the coast. It was, moreover, clear that in view of the marketing possibilities, first consideration must be given to supplying one of the largest steel-using branches of Netherlands industry, shipbuilding. For this the open-hearth process was the only suitable one. For this reason we had no hesitation in deciding that the rolling programme should be concentrated on the production of plates suited primarily for the ship-building industry. Because of the limited financial resources it was further decided that, as a starting point, the capacity of the new rolling mill should be carefully tuned to the needs of the Netherlands market. In choosing the width of the heavy plate mill, which was fixed at 114 in. (net width of plate 100 in.), we aimed, as a result of an extensive enquiry among future consumers, at reaching an equilibrium between the higher costs of a wider mill as compared with the anticipated higher returns from wider plates and the more frequent demand for them.

Construction of the steelworks began in 1938, and in succession four 80-ton furnaces were completed. In connection with the construction of the rolling mill, which was started in 1939, it was important to decide how best to provide for the quality that would be demanded in the plate with respect to both surface and thickness tolerances. To comply with these demands, it was decided to adopt the two-high four-high tandem arrangement, even though the initially estimated production did not make this choice necessary. We were, I believe, the third steel producer in the world to choose such an arrangement, although the Appleby-Frodingham Steel Company was setting up a similar plant at the time. In any case only the Lukens Steel Company in the U.S.A. and the South Durham Steel and Iron Co., Ltd., preceded us. Our rolling mill, however, was wider and heavier than either of these.

The annual production anticipated when the mill was put into operation was 120,000 tons, but the capacity has been increased several times in the course of the years, and will shortly reach between 350,000 and 400,000 tons a year. The maximum plate weight then will be $9\frac{1}{2}$ tons and the maximum net width 110 in. Although, therefore, the rolling mill has been satisfactorily adapted to the increased marketing possibilities and, as regards equipment and construction, is still sufficiently modern and operating at a low cost, the width of the plates as well as their weight will probably become insufficient in time if the



Black = 1924, grey = present layout, dotted line = steelworks and rolling mills as planned 1920

Fig. 9.—Layout of IJmuiden works

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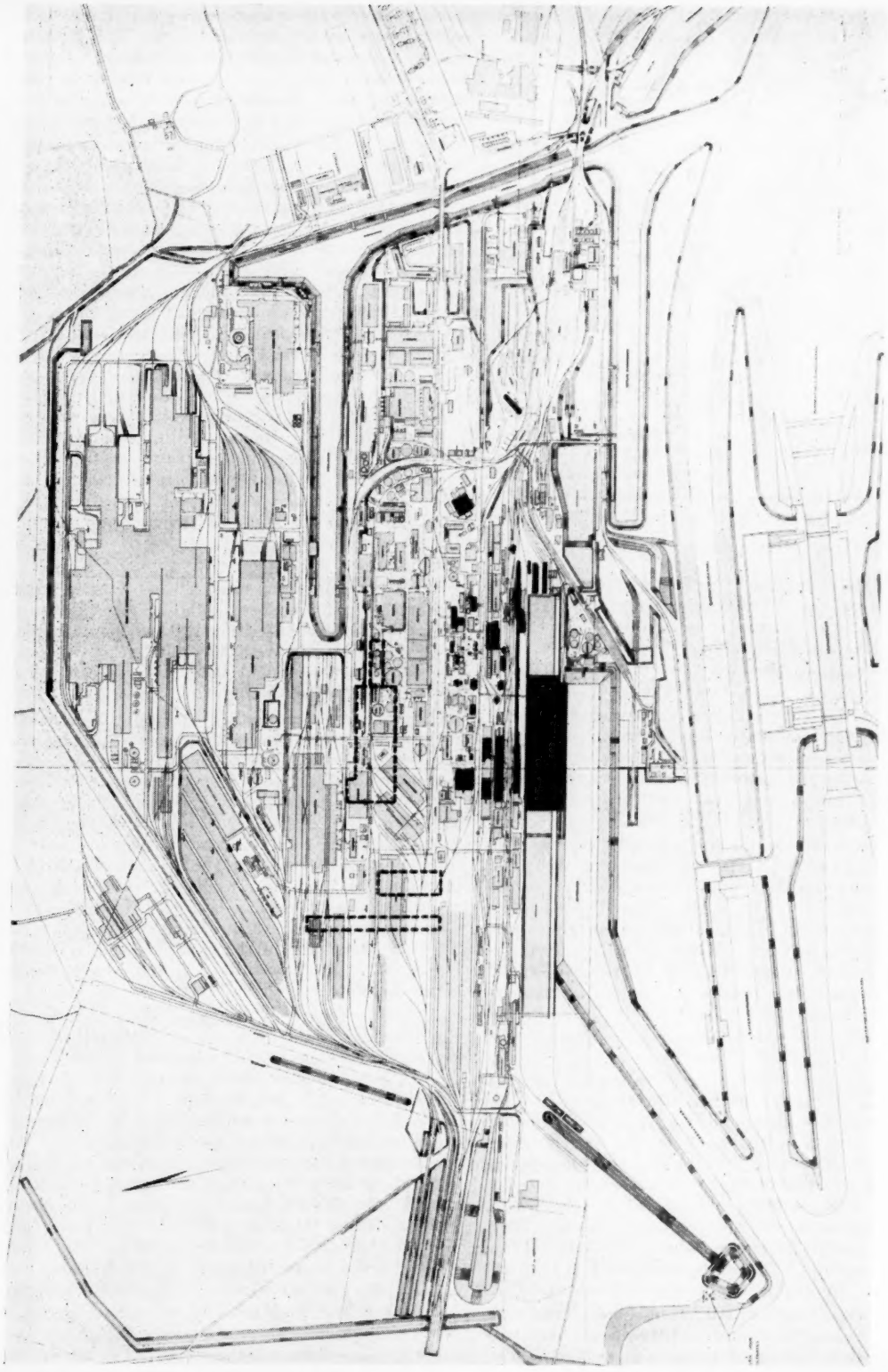

Black = 1924, grey = present layout, dotted line = steelworks and rolling mills as planned 1920

Fig. 9.—Layout of IJmuiden works

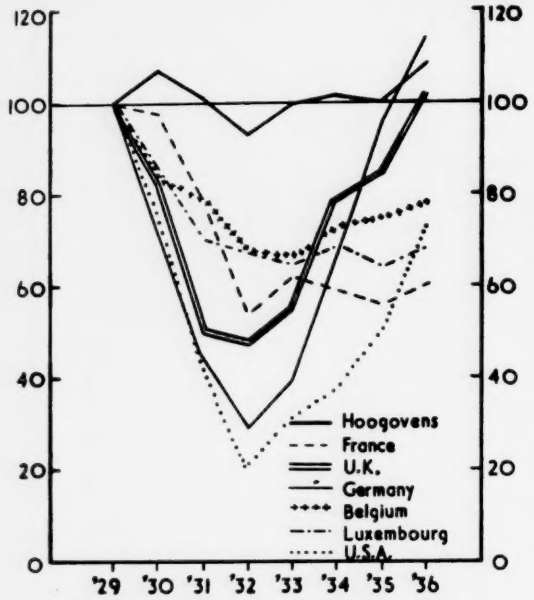

Fig. 10—Pig-iron production (1929 = 100)

size of ships continues to grow. So far as I know, several producers of plates in the U.K. also are for this reason in the course of broadening their rolling mills or are considering doing so.

In the entirely new post-war situation our full attention was first focused on attaining a production as large as possible with the existing installations. I have already mentioned the increase in production achieved in the blast-furnace section to about 700,000 tons a year by the introduction of continuous three-furnace operation and the expansion of internal transport capacity which it entailed.

Expansion of steel production was achieved by constructing two additional open-hearth furnaces, and by enlarging the four existing 80-ton furnaces to 200-ton furnaces. This, among other things, meant a changeover to oil firing. These two decisions were interconnected. As the change to oil rendered the gas regenerators superfluous, it was possible to expand furnace capacity to a maximum and to accommodate the required regenerator volume, within the same space between the existing columns. Figure 11 shows how this was done.

Heat consumption averages about 4 million B.t.u./ton of ingots, whilst 900,000 B.t.u./ton are reclaimed in waste-heate boilers. The desire to make the maximum use of the heat in the steel furnaces led in 1947 to research into the principles underlying the burning of oil under different conditions, initially, of course, with limited means. Gradually this fundamental research, initiated entirely at IJmuiden, attracted great international attention, and I am very pleased to be able to mention that the first country to co-operate in this research was the U.K.

The scope of this research has now been broadened, not only because the number of participating countries has increased but also because the problem under investigation has extended beyond the specific requirements of the steel industry. The following countries are now taking part in this international flame research: the U.K., France, Belgium–Luxembourg, the U.S.A., and Sweden, whilst the European Coal and Steel Community has made a financial grant.

Another means of achieving the greatest possible production with the existing furnaces which was adopted was conversion to all-basic furnaces. In this way, not only was the hourly production increased, but also the duration of the campaign could be raised from about 100 to 200 heats. The high consumption of basic refractories and their high cost, however, justify an intensive study of the material itself and of the factors which influence its consumption. In the U.K., co-operation in this field was sought from BISRA and the British Ceramic Research Association. The All-basic Furnace Sub-committee which was born from this co-operation has already investigated many aspects of this extraordinarily difficult problem. In this connection I would record here our gratitude for the fact that some members of the staff of Hoogovens at IJmuiden are regularly invited as the only foreign guests to the meetings of this Committee, and I am happy that they are given the opportunity to contribute from our side to the development of this important research.

Recently, important speeding-up of open-hearth steel production has been obtained by the use of desiliconization, a pre-treatment which is now applied to all molten iron. With 50% hot metal, the heat-time is reduced by about 10%, which means about 1 h for the 200-ton charge.

Of course, it was not only by enlarging the capacity of the existing installations that the post-war increase of production was achieved. Many new installations have been added to the works in the course of the past ten years. The first and most important of these was a wide-strip mill. The U.K. had led the way before the war in adopting this American development, but on the Continent the wide strip mill was not introduced until after 1945. True, a rolling mill of this type, but of limited proportions had been built in Dinslaken in Germany before the war, but there, in my opinion, too little use was made of the experience gained in the U.S.A., so that the results achieved were not altogether satisfactory.

A thorough study of the possibilities of extending the IJmuiden works led to the conviction that under post-war conditions as well, flat-rolled products would be the most obvious line to adopt. Our thoughts therefore turned in the direction of sheet production. For a short time the choice between a Steckel hot mill and a semi-continuous mill was under discussion. The former meant lower investment costs and would have sufficed for the production of the tonnage which was initially anticipated, but as our concern has always been guided by the desire only to supply products of the very best quality, definite preference was finally given to the semi-continuous mill. Moreover, although the expected sales on the Netherlands market were taken as a basis initially, the future growth of the consumption of sheets had, from the outset, been taken into account. On these grounds it was decided

to build a semi-continuous mill. It was not the intention to manufacture the wider types of sheet required by the automobile industry; on the contrary, it was decided that tinplate, the demand for which in our home market is comparatively great, should constitute an important part of the production programme.

In connection with this rolling programme and on account of the desire to install a combination tandem cold mill instead of several reversing cold mills, the width of the hot strip mill was deliberately limited to 56 in. The minimum number of four hot 'finishers' with which we started has since been brought up to six, the present capacity being 650,000 tons of coils a year. This capacity can, however, easily be doubled. We now know that a semi-continuous mill is only a little more expensive with regard to conversion costs than a continuous mill working at full capacity.

The cold mill, the cleaning lines, and the annealing system do not need any special comment, as they are all of the conventional type. With regard to the tinning process, we naturally had to conform to the needs of our home market. The food processing industry has special requirements, long experience, and risks. It demands principally hot-dip material. The development of the application of electrolytic tinplate is not nearly as general in Europe as it is in the United States.

Home consumption is strongly tied up with the habits of the Dutch housewife, who regards a tinned can more as a matter of investment than as an item of daily use. In addition, the export of foodstuffs is mainly directed to the tropics with their hot damp atmosphere.

Nevertheless, there is also a trend towards electrolytic tinplate in Holland, and for that reason we are building a Ferrostan electrolytic tinning line. It is certain, too, that the consumption of cans in Europe will increase significantly so that the saving in tin costs will also have an important bearing on the

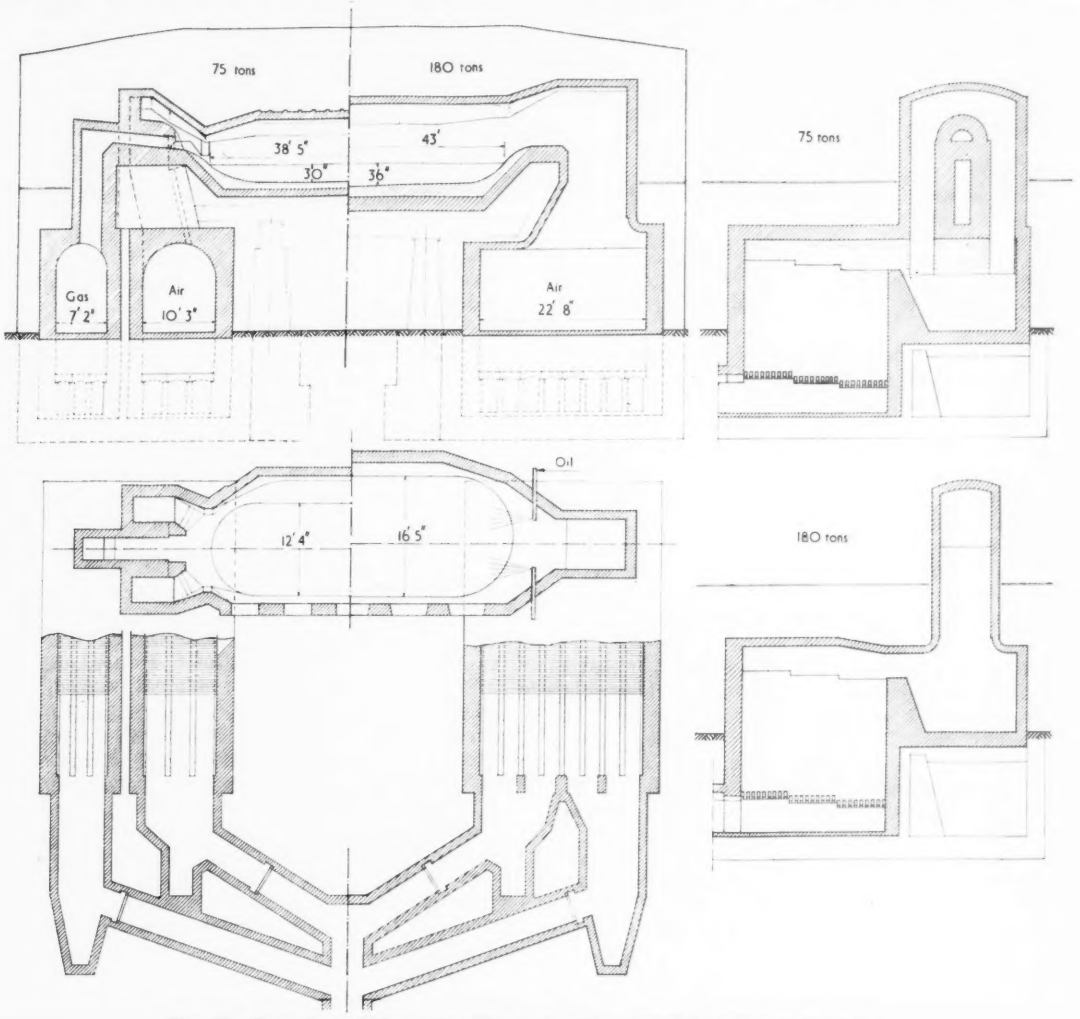

Fig. 11—Open-hearth furnaces; 75-ton furnace rebuilt as 180-ton furnace

Fig. 12—No. 4 blast-furnace

expansion of this market. Moreover, we expect that in time many beverages will be sold in electrolytically tinned cans.

In my opinion it is a problem of general economics whether the standard of living in our countries will so rise that our housewives will be able to afford more prepared canned food as the number of domestic workers decreases, and as there is increasing need for easier household management.

Considering all these factors, we as manufacturers of tinplate must, however, always pay close attention to the development of other packaging materials. In this respect, the use of steel sheet will, in my opinion, always play a very important role, because its potential appeal will gradually cover an ever wider field.

I have already drawn attention to the fact that as well as the enlargement of the capacity of the existing plant, important extensions were carried out. The steelworks was enlarged by the construction of two new 200-ton open-hearth steel furnaces. Three new coke-oven batteries have been added to the existing three since the war. Figure 12 shows a view of the new 28-ft dia. blast-furnace, which is being built. I should also like to show that our new coking-coal stockyard, with a capacity of 150,000 tons, is not served by cranes, but by bulldozers (Fig. 13).

In the manufacture of coke the pre-war method has been followed in that all coke ovens are fired with blast-furnace gas and most of the metallurgical furnaces with oil or blast-furnace gas; the coke-oven gas produced is used for the manufacture of nitrogenous fertilizers or sold to neighbouring cities, including Amsterdam.

I might mention in passing that in the initial period persuading municipalities to take our gas was far from easy, because the larger towns in particular wished to have a wholly independent gasworks of their own. But in the Netherlands since the war there has been a strong move towards centralization of the gas supply in a national grid. In a few years the number of a local gasworks will be reduced, and only a few will remain in operation. The gas output of all the others will have been replaced by 'long-distance gas' which at the moment supplies 40–50% of total public requirements. Coke-oven gas from the steel industry in the north-west and from the State coal-mines in the south, natural gas from the east, and gas from the large oil refineries near Rotterdam will shortly be combined in a great national gas grid.

To effect the coupling of the various networks, it is planned to build a distribution centre in the middle of Holland to permit the exchange of high-calorific value natural gas, refinery gas, and coke-oven gas. The link-up of the various producers concerned and of the numerous municipal consumers in this common grid will be brought about by voluntary private co-operation, with government help. It may be interesting to mention that a national gas grid also renders possible an international exchange of gas.

The supply of gas by our works to towns amounts to more than 7,000,000,000 ft^3 (200 million m^3), of which half is taken by Amsterdam, the capital.

In 1954 the favourable market for rolled products

Fig. 13—Stockyard for coking coal

and the large steel rolling capacity once more brought under review the question of how to extend our steel production still further. This need for greater steel capacity once again focused attention on the choice of the manufacturing process. This choice was now no longer as self-evident as it had been 20 years before. I have already pointed out that, in my opinion, one of the principal characteristics of the period 1910–1950 had been the comparatively static nature of iron and steelmaking processes compared with the developments in rolling which took place in that same period. In the latter, for example, we saw, apart from the revolution in the rolling equipment (the introduction of continuous rolling), the development of cold rolling and drawing practices, and the various methods of surface treatment and annealing.

In steelmaking the choice for large integrated works lay for 80 years essentially between open-hearth and basic Bessemer. However, for some years now metallurgy has been making great strides. In steelmaking this has already been shown by the practical application of new processes which constitute an intermediate stage between the open-hearth and basic Bessemer processes. I can only mention here improved converter steel, oxygen converter steel originated by Linz-Donawitz, the rotary kiln converter, and possibly also in time 'Cyclo-steel,' some advance details of which have already been revealed by Sir Charles Goodeve.

Extensive study of the oxygen converter steel process has shown that it has a number of technical

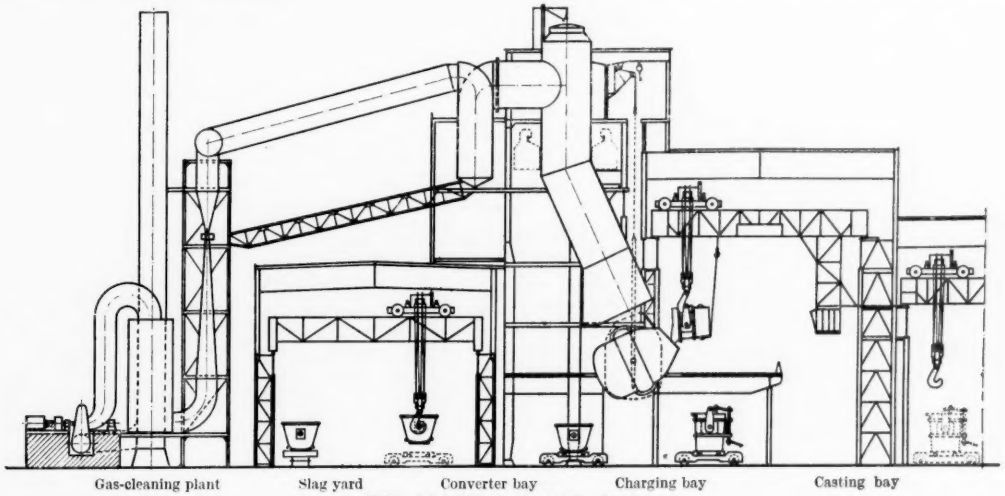

Fig. 14—Oxygen steel plant

and economic advantages. In particular, the small scrap consumption is an extremely important factor in view of the ever-growing difficulty of providing for the scrap needs of the European steel industry. The disadvantages of operating two different steelworks have proved to be comparatively small, because among other things the same pig-iron composition is suitable for both open-hearth and oxygen steel. Nor does the large demand for oxygen constitute an objection, because the presence of a nitrogenous fertilizer plant means that the production of nitrogen for this factory can be combined with oxygen production for steelmaking. Moreover, the addition to this factory of an installation for the partial combustion of fuel oil for the production of synthetic gas has created there, too, a greater need for oxygen. For these reasons we have in IJmuiden an excellent basis for the manufacture of tonnage oxygen. When, moreover, a study of the quality of this steel showed that, with proper precautions, it can be considered equal to open-hearth steel, it was decided to build a new works for the oxygen steel process with two top-blown converters (Fig. 14). This will start production at the beginning of 1958 with an initial capacity estimated at 300,000 tons a year.

This is another example of parallel developments in the steel industry, for in the U.K. also great interest is shown in new processes. I may mention, for example, the basic-Bessemer process with oxygen-steam-enriched blast which is to be used by The Steel Company of Wales Ltd.

The development in iron- and steelmaking technique, reviewed only summarily here, together with the changes in the rolling, annealing, and other processing methods, affect the way in which the quality of steel products must be determined as, for example, for standardization purposes. The progress outlined has made it possible not only to improve quality and to exercise better control over given qualities, but also to manufacture finished products with almost identical properties by different processes. Where previously it had seemed adequate in many cases to determine the quality approximately by specifying the manufacturing process, this is now no longer justifiable either technically or economically.

Then, too, the techniques of inspecting and testing the final products have been improved, and as a result there is a better understanding, both by consumers and producers, of the relation between test results and the properties required in the finished product. From this has developed a tendency to prefer the determination of quality by means of test values rather than to specification of the manufacturing process.

It is in the interests of the steel producer as well as of the consumer that the standards with regard to rolling products should be adapted to this development. Many institutions are now working to this end; I should particularly like to mention Lloyd's Register of Shipping, which is fully alive to the changes that are taking place and is in the forefront in revising classification methods. But the other shipbuilding classification societies, the International Institute of Welding and the International Organisation for Standardisation, are also working on this problem. One of the sections of the High Authority of the European Coal and Steel Community is also engaged on this matter.

But surely there is a danger here in that too many international organizations are tackling the same problem. The association between the U.K. and the European Coal and Steel Community, which is at present, in my opinion, assigned too little useful work, might well assist in co-ordinating the various activities in the field of standardization and classification, thereby avoiding too great a variety of views and specifications.

I should like to conclude with a few remarks of a more general character.

We observe in the whole field of action of our iron and steel industry an increasing parallelism in many different forms of development. Political, economic, and technical factors move in the same direction. Viewed politically, the strong isolation and separation of various countries is now slowly becoming less marked, as is shown by the creation of the European Coal and Steel Community and, in very recent weeks, the Treaty for a European Economic Community.

So far as the steel industry on the Continent of Europe and in the U.K. is concerned, time and again we come across the same problems, such as, for example, a fundamental shortage of scrap and of coking coal.

From a technical standpoint, the parallelism which has resulted from seeking solutions for the same economic problems is being still further promoted by a fact which has always impressed me personally, namely, the great willingness of the steel industry to exchange knowledge and experience. Technical discoveries and improvements are consequently being spread much faster than would otherwise be possible, to the great benefit of our industry as a whole.

Contacts in the field of quality control and of research into the many kinds of application are continually becoming closer and more frequent. Our Iron and Steel Institute plays an important role in this respect, both by the personal international contacts it brings about and by its excellent relations with similar institutions in other countries, as our overseas visits and the very warm welcome we always receive bear excellent testimony. Yet in the field of joint research and similar activities I believe that, by joining hands internationally, we as friendly institutions can promote a still greater number of facets of the development of our industries.

In conclusion I can only say that I sincerely trust that, by letting you in this address look through the eyes of a foreigner, I have been able to help show you the international foundations of our profession, foundations on which all of us, as national industries, side by side and filled with the same spirit, can carry out our national tasks within the great complex of the world iron and steel industry.